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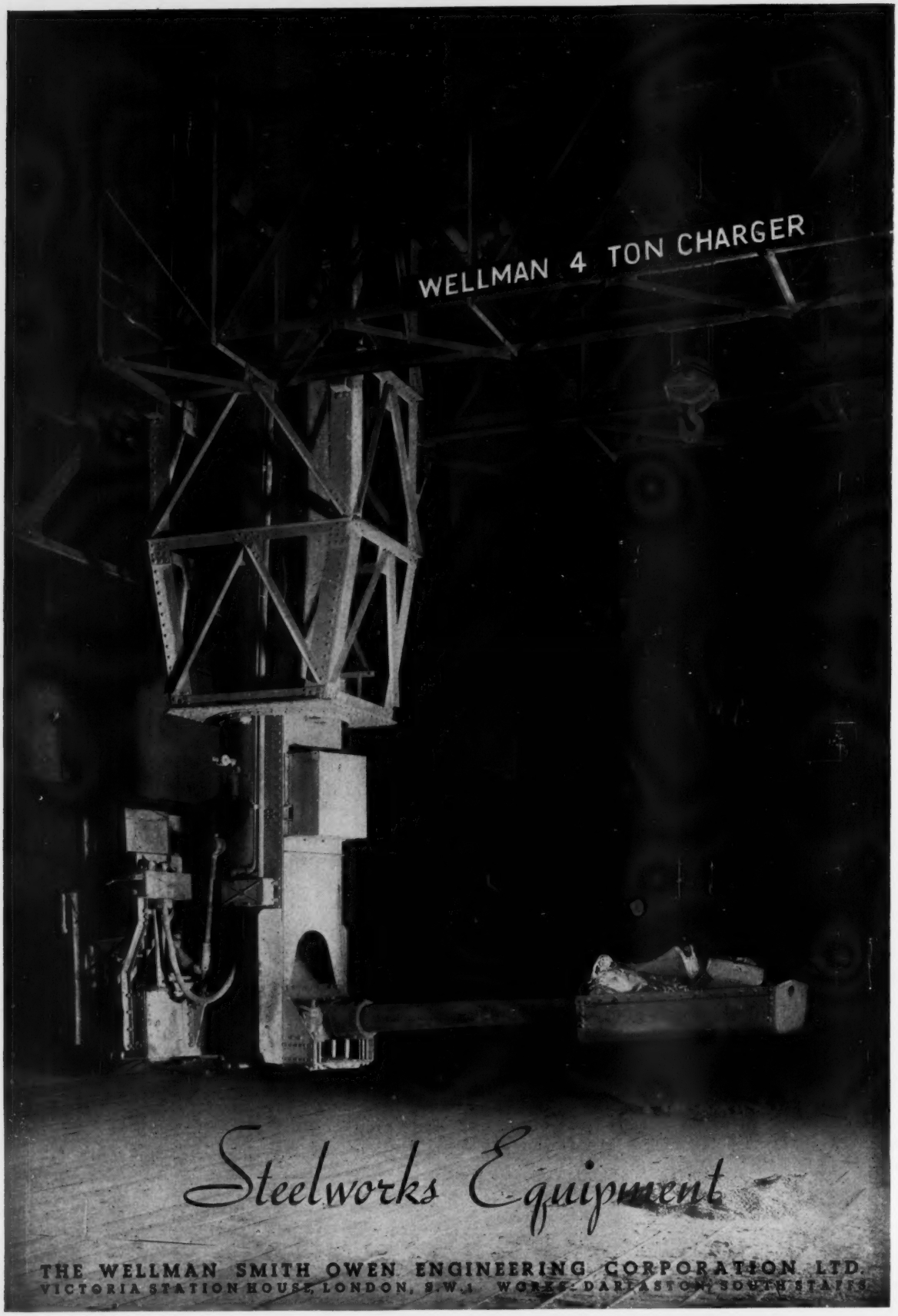
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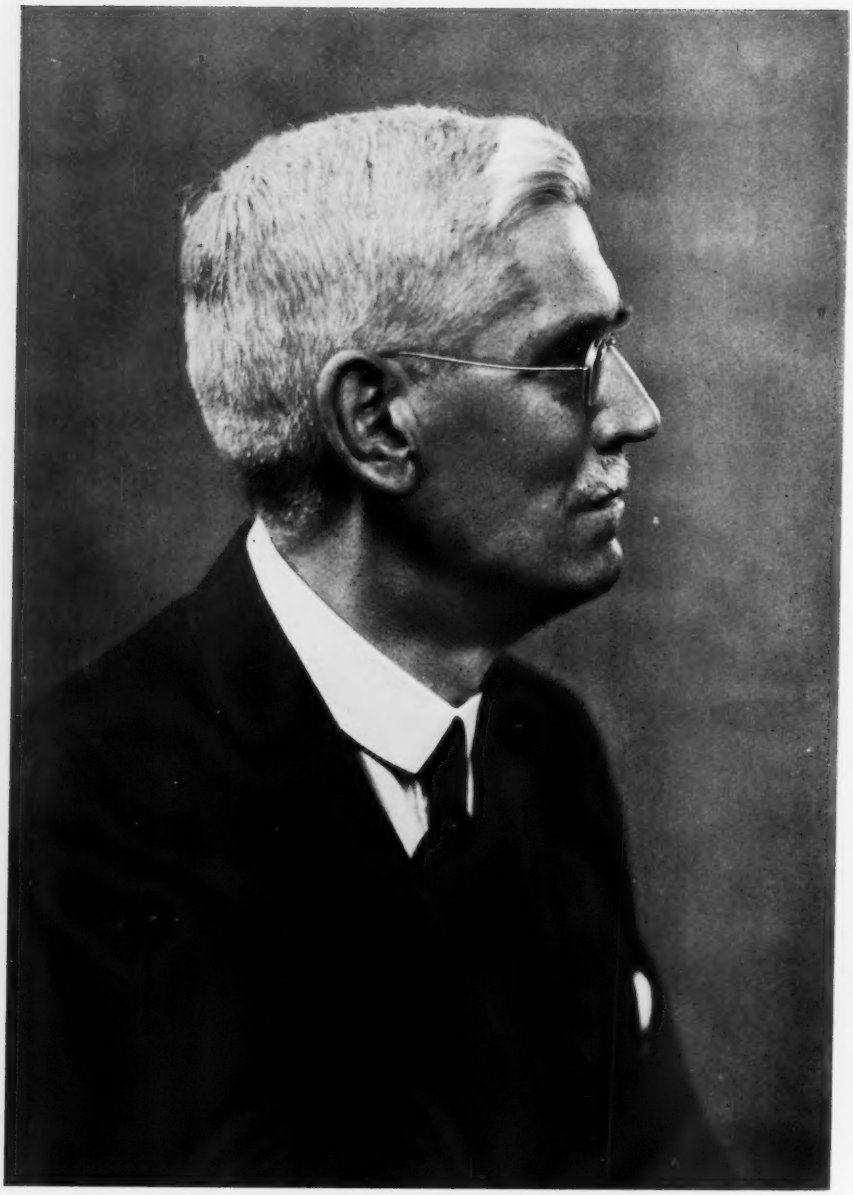
SIR WILLIAM LARKE, K.B.E., D.Sc.

SIR WILLIAM LARKE, K.B.E., D.Sc., received his early engineering training with H. F. Joel & Company and Siemens Bros. & Co., Ltd., Woolwich. In 1898, he joined the British Thomson-Houston Company, and acted as Engineer and Manager of their Power and Mining Department from 1899 to 1912; in 1912, he was appointed Executive Engineer and served in that capacity until 1915. He then joined the Ministry of Munitions and occupied various responsible positions before being appointed Director-General of Raw Materials. He received the O.B.E. in 1917, became C.B.E. in 1920 and was knighted in 1921.

In 1922 he was invited to become the Director of the National Federation of Iron and Steel Manufacturers, which was formed as part of an organised effort by British industry to lift the country out of the depression ruling at that time. It soon became evident that some control of the import position was necessary if industry were to regain its prosperity, and this involved the Federation and its Director in a campaign, designed to impress upon the country and the Government of the day, the soundness of these views. The prosecution of the campaign involved the creation of an organisation competent to produce the economic and other information on which the Federation policy was based. Co-operation within the industry was fostered, in order that its economic interests be protected and its technological efficiency increased. The organisation of co-operative research was undertaken by the Federation in 1924 and the activities of those earlier days have resulted in the present British Iron and Steel Federation and the British Iron and Steel Research Association. The latter carries on the work of the Iron and Steel Industrial Research Council, of which Sir William was Chairman from 1938 to 1945.

Sir William became a member of the Iron and Steel Institute in 1904 and has served for many years as a member of Council and on various Committees of the Council, notably the Finance and General Purposes, Publications and Joint Library Advisory Committees. He was elected Vice-President in 1934, becoming Honorary Vice-President in 1946. He has also been President of the Institute of Fuel, of which he was one of the founders, of the Junior Institution of Engineers, of the Electrical Research Association and of the Institute of Welding. He played a prominent part in the formation of the Welding Research Council in 1935, of which he was Chairman until the formation of the British Welding Research Association, of which he was elected the first President in 1945. He has also been Chairman of the Scientific Instrument Research Association and is at present Chairman of the British Standards Institution and of the Industrial Research Committee of the Federation of British Industries.

His many services to industry and unfailing appreciation of the value of research were recognised by Durham University in the conferment of the honorary degree of Doctor of Science. A further tribute to his services to science and industry is now paid by the Council of the Iron and Steel Institute, who have announced the award to Sir William of the Bessemer Gold Medal for 1947.



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Sir William J. Larke, K.B.E., D.Sc.
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IRON AND STEEL
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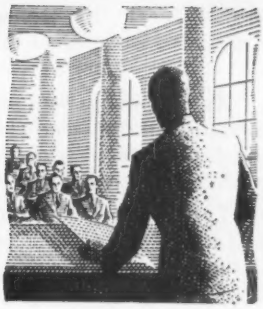

Graphite Formation in Cast Irons and in Nickel–Carbon and Cobalt–Carbon Alloys*

By H. Morrogh† and W. J. Williams, B.Sc.†

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* Received 4th May, 1946. † British Cast Iron Research Association, Birmingham.

Synopsis

An attempt has been made to give a concise statement of as many as possible of the features of graphite formation in cast irons which need explanation, together with a critical examination of the numerous theories which have been developed from time to time. A detailed description of the metallography of graphite in relation to its crystal structure is given, and the solidification of flake-graphite-containing irons is described.

The process of undercooling in cast irons is examined with reference to a number of problematical points and several new phenomena with respect to graphite formation in cast irons have been introduced. The effects of titanium and zirconium in refining the graphite depend upon the presence of the carbides of these elements. Two types of inverse chill have been demonstrated.

The process of graphite formation in nickel–carbon, nickel–iron–carbon, and cobalt–carbon alloys has been studied, and the analogies between the mechanism of the process in these alloys with that in cast irons have been worked out. Undercooling in all these alloys and in cast irons takes place by the solidification of a eutectic of solid solution + carbide, and the carbide phase decomposes after solidification. Coarse graphite structures solidify as a eutectic of solid solution + graphite. Nickel–carbon and cobalt–carbon alloys tend to give coarse graphite when they contain more than a certain amount of sulphur. In the absence of sulphur additions, spherulitic nodular graphite structures may be produced in nickel–carbon, nickel-rich nickel–iron–carbon, and cobalt–carbon alloys by the addition of calcium and/or magnesium in a wide variety of forms. The manner in which these spherulites form has been studied.

The formation of spherulitic nodular graphite in cast irons without heat-treatment has been shown to occur. The conditions necessary for this have been outlined, and spherulites have been produced artificially without heat-treatment.

The undercooling of cast irons and of similar alloys has been discussed at length, and the more general aspects of the particular type of undercooling have been emphasized.

Introduction

The presence of graphite in grey cast iron, and to a lesser extent malleable cast iron, confers upon the material its characteristic properties. The relatively low tensile strength and poor shock resistance on the one hand and the good damping and free-machining properties on the other are largely determined in grey cast iron by the presence of graphite. By modifying the amount and shape of the graphite the properties of the material can be considerably changed. The mode of occurrence of graphite in various cast irons and pig irons is known to vary in a manner not always explicable in terms of the known facts or current theory. These variations have very important practical and commercial repercussions. Normally, a grey cast iron has graphite in a lamellar form of some type, but it is well known that if the graphite phase is in the nodular form, as in a heat-treated malleable iron, a material of improved shock resistance and ductility is obtained. In this paper it is the prime concern of the authors to deal with the formation of graphite during the process of solidification and subsequent cooling of castings, and not with the formation of graphite during a heat-treatment process, although, as will subsequently be shown, the two are not disconnected. It is the authors' opinion that there can be produced in cast irons a wide variety of graphite formations which have not yet been reported, still less explored and commercially produced, and which make imperative important changes in the theory of graphite formation.

Until the mechanism of graphite formation in cast iron is fully understood, the full metallurgical potentialities of the material will not be realized. This process presents problems as difficult of solution as any in the whole field of metallurgy. Cast iron is one of the most complex alloys used, and its intrinsic properties are such that experiment is very difficult. There is a very large amount of literature on the subject, but there are still broad gaps in our knowledge. In what follows no attempt will be made to review the literature in any great detail, and reference will be made only to what, in the authors' opinion, is the most recent and most authoritative work. However, an attempt will be made to summarize the existing knowledge of the mechanism of graphite formation and to state the problems and phenomena which need to be explained and linked together. This will be followed by an account of the authors' experimental work; this is by no means complete, and at the present time it appears that the various problems involved require careful experiment which can only be carried out as part of a long-term research programme. The work described here represents the completion of a stage of such a programme.

Much of the experimental work described in this paper deals with the formation of graphite in nickel–carbon and cobalt–carbon alloys. It is

not immediately apparent that this is related to graphite formation in cast iron, but later in the paper reasons are given for the conduct of this investigation and its presentation in this context.

A survey of the literature on the subject reveals a very confusing state of affairs, with numerous theories in existence which in many cases have resulted from work carried out under special experimental conditions, not necessarily generally valid. Attempts to generalize from such theories have involved some contradictions. Before dealing, however, with the debatable issues, the points which are generally accepted as true will be stated. It is significant that such a statement can be devoted only to the broadest issues.

It is well known that cast irons are essentially iron–carbon–silicon alloys containing also varying amounts of manganese, sulphur, and phosphorus as an invariable rule, and that frequently nickel, chromium, copper, molybdenum, titanium, vanadium, arsenic, etc., are present singly and in various combinations. When considering cast irons it is convenient to refer them to the iron–carbon–silicon ternary system, but cast irons differ in many significant ways from alloys of this system. This difference must always be borne in mind. For the purpose of the subject under consideration it can be said that the equilibrium diagrams for the binary iron/iron-carbide and the iron–silicon systems are known with fair accuracy. The ternary iron–carbon–silicon system under equilibrium conditions is much less clearly defined and presents opportunities for considerable further investigation. The issue is, however, complicated by the fact that under practical conditions the iron–carbon–silicon alloys exhibit a metastability anomaly. Under some conditions of composition and cooling, steels and cast irons have major microstructural phases consisting of iron carbide (cementite) and a solid solution of carbon in α or γ iron. In these cases the carbon is said to be in the combined form. On the other hand, according to its composition and thermal treatment, a steel or cast iron may have major microstructural phases consisting of graphite and a solid solution of carbon in α or γ iron. It is generally assumed that iron carbide (cementite) can be considered as a metastable, or thermodynamically unstable, compound of considerable persistence under certain thermal conditions and compositions. This has led to the postulation of the existence of two similar systems, one the metastable system, in which the carbon not in solid solution is present as cementite, and the other the stable system, in which the carbon not in solid solution is present as graphite. In both cases the carbon is assumed to be in the same condition in the liquid state.

In the case of cast irons the situation is further complicated by virtue of the fact that, during cooling, transformations at different stages in the cooling may occur along metastable or upon stable lines. For instance, solidification may proceed in accordance with the stable system, giving, immediately after solidification, a structure of graphite in a matrix of austenite, but this austenite can subsequently transform to pearlite according to the metastable system, giving as a result the familiar pearlitic grey cast iron. Similarly, solidification may occur partly according to the stable system and partly according to the metastable system, giving a mottled iron.

As a general rule, rapid cooling tends to give phases predicted by the metastable system and slow cooling tends to give phases as predicted by the stable system. Alloying elements, *e.g.*, sulphur, chromium, manganese, vanadium, molybdenum, tellurium, etc., may favour the transformations during cooling and solidification to give metastable products. On the other hand, other elements may enable transformations to occur according to the stable system, for instance silicon, nickel, etc. The reason for this behaviour of alloying elements in cast irons is not yet understood. As far as carbide stability is concerned, elements such as cobalt, titanium, and zirconium may have little effect. The proximity of microstructural graphite to cementite produces further instability, the effect becoming more pronounced the finer the graphite particles. Hydrogen gas and hydrogen in the form of water vapour may, in contact with a melt, give increased carbide stability. In a less easily defined manner it is probable that oxygen also influences carbide stability. The tendency for phase changes to occur according to the stable system is greater at higher than at lower temperatures.

In addition to the dual behaviour of cast irons referred to above, a further complication exists. When a cast iron solidifies in the grey condition it may do so in one of two ways, or in a wide range of combinations of these two ways. The graphite which forms at or near to the minimum solidification temperature—the " eutectic " temperature—may occur in a lamellar form commonly known as flake graphite, or in a finely divided form variously referred to in the literature as supercooled, reticular, undercooled, eutectic-like, dendritic, etc., graphite. At this stage it is not proposed to illustrate the differences between these two structures micrographically. The literature contains many illustrations of the differences, and subsequently in this paper the differences will be

dealt with and illustrated in some detail. The mechanical properties and general behaviour of a cast iron are considerably affected by the type of graphite present. In irons which are cast white and subsequently annealed to decompose the carbide, the resulting graphite is generally in a nodular form. In this form the graphite does not exert such a profound embrittling effect as either of the varieties found in normal grey cast irons and referred to above. The production of nodular graphite forms the basis of the method for manufacturing malleable cast iron. It is generally assumed that the nodular form of graphite can be formed only by the heat-treatment of white cast iron or steel. One of the objects of this paper is to indicate that it is possible to obtain nodular graphite structures in cast irons without heat-treatment.

In what follows, the important anomalies relating to carbide stability and graphite formation will first be discussed.

Of the micrographs reproduced in this paper, Figs. 16–19, 23–25, 29, 31–56, 58, 59, 62, 74–78, 82–84, 86, 87, 90, 92, 94–97, 99, 101–104, 106–109, and 113–123 were taken with a 16-mm. achromatic objective, N.A. 0·28, without an eye-piece. A 16-mm. apochromatic objective, N.A. 0·30, was used for Figs. 9–12, 20, 21, 64, 65, 79, and 81. A 12-mm. achromatic objective, N.A. 0·50, was used for Fig. 68. A 4-mm. achromatic objective, N.A. 0·83, was used for Figs. 13–15. A 4-mm. apochromatic objective, N.A. 0·95, was used for Figs. 66, 110, 124, and 125. A 2-mm. oil-immersion achromatic objective, N.A. 1·3, was used for Figs. 57, 60, 61, 63, 67, 69–73, 85, 88, 89, 91, 93, 98, 100, 105, 111, and 112.

Crystal Structure of Graphite and its Metallography

In many ways graphite is a unique phase in ferrous alloys (and in non-ferrous alloys, with the possible rare exceptions of compounds such as calcium silicide ($CaSi_2$) and molybdenum sulphide (MoS), which have similar layered hexagonal structures). It has a crystal structure differing considerably from those of the other phases present, and this fact no doubt is responsible for many of its peculiar properties. Graphite in cast irons has a structure similar to natural and synthetic graphites,[1,2] which is layered hexagonal (Fig. 1).

When considering the growth of graphite crystallites in cast irons, it is desirable to keep this structure in mind. The carbon atoms lie in a series of parallel sheets, in each of which they are arranged at the corners of a set of regular hexagons. The sheets are so superposed that only

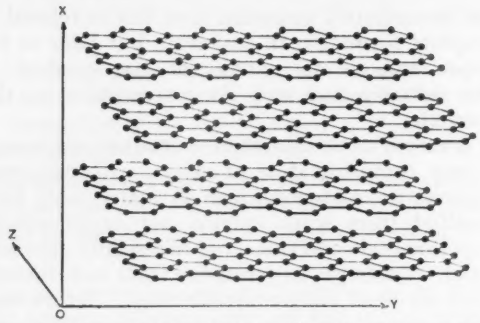

Fig. 1—The graphite lattice. Growth of a graphite flake occurs preferentially in the plane *OZ*, *OY*

one-half of the atoms in each sheet lie vertically above atoms in the sheet below. Alternate layers lie, atom for atom, normally one above the other. The carbon atoms in the layers are very closely spaced compared with the spacing between adjacent planes. This permits easy slip along these planes and accounts for the characteristic mechanical properties of graphite. Recently the presence of another graphite structure in small amounts in both natural and kish graphite has been described.[3,4,5] The metallurgical significance of this has not yet been stated.

Graphite flakes in cast iron of both the "normal" variety and the "supercooled" variety are essentially basal sections of graphite. With the exception of kish graphite, the flakes are generally distorted in all directions and may be spatially arranged in complex shapes and groupings.[6,7] In general, however, a graphite flake which appears on a polished microspecimen is so oriented that the basal planes meet the polished surface in a direction approximately parallel to the visual length of the flake if each graphite flake is considered as a single crystal. In other words, the longest dimension of a graphite flake is approximately at right angles to the direction *OX* (Fig. 1).

As the polished surface of a microspecimen of grey cast iron represents a random section through a large number of randomly oriented graphite flakes, there will be a statistical deviation from the orientation relationship given above; hence its approximate character. The probability of a graphite flake occurring with its basal plane parallel to the polished surface exists, but is very slight. When such a case occurs it is impossible to polish the graphite, owing to slip giving exfoliation.

In sections parallel to the basal planes, graphite has a metallic appearance with a high reflectivity and does not exhibit any form of optical anisotropy. Nipper[8,9] has shown photographs of

basal sections of kish graphite lamellæ presumably scraped from the fractured surface of a sample or from the top surface of a casting head. These clearly reveal the properties referred to above and also show hexagonal markings, as would be expected from the grouping of the atoms in the basal planes. Sections of this type are never seen in polished microspecimens. The sections of graphite flakes normally visible in cast iron are easily polished, have a non-metallic appearance, and exhibit pronounced optical anisotropy and reflex pleochroism.[10] Such polished graphite sections have a light grey-brown colour in unpolarized light, and light grey-brown to black in plane- or partially plane-polarized light, the actual appearance of any graphite crystallite being determined by the reflex pleochroic effect. No phase in cast iron should be termed graphite unless it has been polished and its optical anisotropy and reflex pleochroism confirmed. In the work described in this paper this was carried out as routine in all cases.

So far, it has been assumed that graphite flakes are single crystallites, but in the majority of cases examination at high magnifications under plane-polarized light and between crossed nicols reveals a complex structure which indicates that graphite flakes are composed of a main crystallite with many smaller and differently oriented secondary crystallites in the interior.[1,10,11] It will be shown later in this paper that these internal graphite structures are not peculiar to cast irons, but also occur in nickel–carbon and cobalt–carbon alloys. Hanemann and Schrader[11] have attributed these internal structures in graphite flakes to mechanical distortion during crystallization. The flat graphite crystals are very pliable. The larger crystals of hypereutectic or "kish" graphite freely floating in the melt sometimes form straight lamellæ, whereas the graphite which forms simultaneously with austenite—the "eutectic" graphite—is restrained during solidification and may be distorted by the melt. Mechanical stressing of graphite crystals may give kinking.[12] It is, for instance, well known[13,15] that when long single crystals of hexagonal metal are grown with the glide plane (0001) parallel to the wire axes, the orientation does not favour the usual glide mechanism under tensile or bending loads. When compressive loads are applied the crystals collapse, forming kinks. The arrangement of the kinking in such metal crystals closely resembles the "zigzag" patterns found in a large number of graphite flakes. It has not been possible to prove or disprove the suggestion put forward by Hanemann and Schrader, but several points can be made which are relevant to the issue. The zigzag formations occur most frequently at bends in graphite flakes, which tends to confirm the mechanical-distortion hypothesis. On the other hand, complex structures not apparently similar to kinking patterns are found throughout some graphite flakes,[10] including perfectly straight "kish" graphite flakes. Again, the zigzag structures are found in portions of graphite flakes not extending through the whole cross-section. For instance, Fig. 9 illustrates an extremely large flake of kish graphite in a blast-furnace-bear sample* viewed under plane-polarized light in its position of minimum pleochroic absorption. The flake will be seen to be substantially free from markings spreading across the whole of its section, but in the regions marked off the zigzag structure is well developed, though on a small scale. This is more clearly revealed in Fig. 10, which shows the same spot between crossed nicols with the graphite flake in its position of maximum brightness. Figs. 11 and 12, which show the same spot between crossed nicols with the graphite flake parallel and at right angles, respectively, to the plane of polarization of the incident light, show the flake to be made up of a number of banded layers. Layers of this type have been shown to occur as the result of the deposition of hypereutectoid and eutectoid graphite on the already existing graphite flakes.[10] However, the zigzag structure appears at the interface between these layers. Figs. 13 and 14 show the area marked off with a full line in Fig. 9 in greater detail at a higher magnification. It seems strange that if these markings are due to mechanical distortion they should not extend across the whole section. Other flakes of kish graphite in the same sample did show markings, each layer having a group of its own, the size of the markings depending upon the width of the banded layer (Fig. 15).

In the case of nodular structures produced by the heat-treatment of white irons, it has been shown[10] that two types exist—the spherulitic and the "graphite-flake-aggregate." The former occurs in irons containing free iron sulphide and the latter in irons containing free manganese sulphide with no iron sulphide. This rule appears to apply for all types of heat-treatment, at temperatures above and below the critical point and with or without pre-quenching. In the course of this paper we shall be largely concerned with the spherulitic type of nodule, and an attempt will be made to show that this structure can be produced in various alloys without any heat-treatment. To the authors' knowledge, the

* Kindly supplied by Dr. J. E. Hurst.

XUM

occurrence of the spherulitic structure in cast irons and in the alloys described in this paper is unique in the whole field of metallography, and may perhaps be attributed to the unique crystal structure of graphite. The structure of graphite spherulites is interesting, and typical spherulites are illustrated in Figs. 58 and 63. In common with other spherulitic structures, graphite spherulites each consist of a large number of acicularly shaped crystallites growing outwards from a centre or nucleus. As a general rule each crystallite is so oriented that its basal planes are at right angles to the radii of the spheroid of which each spherulite can be considered to be composed. This orientation relationship is illustrated schematically in Fig. 2 and can easily be confirmed by examination of a nodule under plane-polarized light. This orientation agrees with the orientation rule for spherulites given by Jansen,[85] who showed that in spherulites the most widely spaced crystallographic planes tend to be oriented at right angles to the spherulite radii. Whilst this is generally true, a few crystallites can be detected in most nodules which do not obey the rule. In whiteheart malleable iron it is frequently possible to find iron sulphide particles at the centres of temper-carbon nodules. When the polished section lies above or below the centre of a spherulitic carbon nodule, the crystallites at the centre of the exposed surface are so oriented that their basal planes lie parallel to the polished surface, and so they cannot be polished. This unpolished zone must not be mistaken for a nucleus.

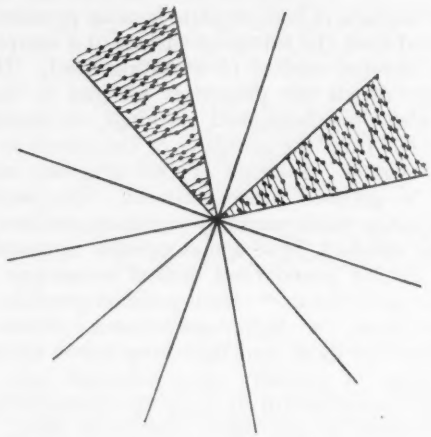

FIG. 2—Schematic representation of the orientation of graphite crystallites in spherulitic carbon nodules

THE SOLIDIFICATION OF CAST IRONS CONTAINING COARSE FLAKE GRAPHITE

Before proceeding further, a description will be given of how, in the authors' opinion, grey cast irons solidify to give flake graphite. The major part of this will perhaps be well known, but many arguable points will necessarily be mentioned. The description has been put in this manner in order to give a defined foundation upon which to erect subsequent arguments. The point to observe is that this description refers solely to irons containing coarse flake graphite, and this will be emphasized again. White irons must necessarily have different treatment, and it is the subsequent purpose of this paper to show that the solidification of undercooled or fine graphite irons is a more complicated process.

The iron–carbon–silicon system shows a trough originating at approximately 1130° C. and 4·3% of carbon in the iron–carbon side of the diagram. With increasing silicon contents this trough moves to higher temperatures and lower carbon contents. In the stable iron–graphite system the point at 1130° C. and 4·3% of carbon represents a eutectic of austenite and graphite. The effect of silicon upon this complex may be represented by the following equation for silicon contents of up to 3%[16]:

$$\text{Carbon, \%, in binary complex} = 4{\cdot}23 - \frac{\text{Silicon, \%}}{3{\cdot}2}$$

In iron–carbon–silicon alloys solidifying in the stable system, this binary complex consisting of flake graphite and austenite has an appearance very different from that of most eutectics, and this has caused many workers to suggest that it does not originate at this type of phase change. However, thermal analysis shows that this phase change does take place at approximately a constant temperature and that it possesses many characteristics of a true eutectic. In agreement with the phase rule, most equilibrium diagrams indicate the transformation liquid $\longrightarrow$ graphite + austenite to take place over a range of temperature. Whether this is so is, however, a doubtful point. Equally doubtful is whether the transformation takes place at constant temperature, as the process is generally accompanied by undercooling phenomena which tend to mask the true thermal effects. For all practical purposes it can be considered as an isothermal transformation.

In most discussions[17–19] on the iron–carbon–silicon system, and in almost all attempts to draw it, it is assumed that carbon enters the system as a carbide. For iron–carbon–silicon systems yielding flake graphite as distinct from undercooled graphite, this is most probably incorrect. Boyles[20] has shown that flake graphite forms directly from the melt, and this has been amply confirmed by Eash[21]; the present authors' experience is also in agreement with this. Jass

and Hanemann,[16] in their construction of the ternary system, have assumed the eutectic to be graphite–austenite.

In most of the literature on the subject the binary complex graphite–austenite is referred to and dealt with as a binary eutectic. The justification for the term eutectic is arguable, but it should be remembered that in cast irons, which in a condition of least complexity are Fe–Si–C–P–Mn alloys, the binary complex can theoretically form over a range of temperature and with changing composition. It is perhaps fortunate that these variables appear to be effective to only an inappreciable degree. If this is thoroughly understood it is permissible to refer to the austenite–graphite binary complex as the stable eutectic and, in the metastable system, to the austenite–cementite binary complex as the metastable eutectic. This terminology will be adhered to throughout this paper and is in line with conventional procedure. Alloys containing less carbon than 4·23% — (Silicon, %,/3·2) will be termed hypo-eutectic, and alloys with more carbon than indicated by this formula will be termed hypereutectic. In passing, it should be noted that other elements, notably phosphorus and chromium, also influence the amount of carbon in the eutectic, lowering its value. The effect of phosphorus is fairly well known, and it is generally assumed that the carbon content of the eutectic is lowered by 0·3% by 1% of phosphorus.

Hypereutectic irons solidify by the direct formation of graphite from the melt in the form of kish. Owing to its low relative density this phase tends to rise to the surface of the melt, but when entrapped by the metal it generally appears in the microstructure as characteristic long straight flakes. This phase is deposited over a range of temperature, starting at the liquidus surface, until the carbon content of the remaining melt is at the eutectic concentration, when a simultaneous crystallization of graphite and austenite occurs. The form of this eutectic graphite in hypereutectic cast irons has many interesting features which will be dealt with later, but as a general rule it occurs in flake form on a finer scale than the primary kish.

In hypo-eutectic cast irons the first phase to be deposited from the melt is austenite. This austenite, a solid solution of carbon in γ iron, crystallizes from the melt in the form of dendrites. The lengths of these dendrites, for a given wall thickness and composition, vary with the pouring temperature—the higher the pouring temperature the longer the individual dendrite lengths. Under the influence of sharp temperature gradients the dendrites are disposed in a columnar pattern, and with no marked temperature gradients they are equiaxed. The carbon content of this austenite will vary with the silicon content, tending to decrease with increasing silicon. At the eutectic temperature (approximately 1130° C.) and with no silicon, it is generally assumed that austenite can contain a maximum of 1·7% of carbon. The solidus surface for the iron–carbon–silicon stable system is not known with any degree of accuracy, and no quantitative statement can be made. The literature contains several suggestions that the solubility of carbon in austenite in equilibrium with graphite is lower than that in austenite in "equilibrium" with iron carbide. Other elements also have an effect on this solubility. With falling temperature, owing to the deposition of austenite, the carbon content of the remaining liquid increases, and at the eutectic concentration the simultaneous deposition of graphite and austenite occurs.

This eutectic forms with a spheroidal crystallization front. The pattern, the number, and the size of these eutectic cells bear no direct relationship to the same characteristics of the primary austenite dendrites. In cast irons containing phosphorus the phosphorus-rich liquid segregates at the boundaries of these eutectic cells and so the resulting phosphide eutectic (steadite) outlines and indicates these cells. The fact that eutectic crystallization proceeds on a spheroidal crystallization front influences the pattern of the resulting graphite, so that the individual graphite flakes tend to be disposed radially in each cell, giving rise to the familiar cluster or whorl formations. This radial disposition is indicative of but a general orientation trend in any given eutectic cell. All the eutectic cells are allotriomorphic, taking shape as a result of contact with neighbouring cells and the availability of space between the primary dendrites. The perfection of the eutectic cells depends to a large extent upon the free space between the austenite dendrites. In all hypo-eutectic grey irons the graphite forms after and hence between the primary dendrites—all graphite in such irons is interdendritic. In low-carbon cast irons the space between the primary dendrites is small, and hence the eutectic cells tend to be very imperfect spheroids and the graphite flakes have to take up positions more markedly dictated by dendrite disposition than by eutectic cell influence. Since in these low-carbon cast irons the primary dendrites determine the graphite-flake pattern, it is obvious that this will be subject to the same influences as the primary dendrites. With high pouring temperatures in low-carbon irons and long dendrites, large series of similarly oriented graphite flakes are obtained.

XUM

With low pouring temperatures and short equiaxed dendrites, the graphite flakes tend to be randomly distributed. At this stage it is necessary to recall that only stable alloys yielding flake graphite are being considered. With high carbon contents dendrite disposition is relatively unimportant, and on the macro scale the graphite is again randomly oriented in the form of clusters. In high-carbon irons the dendrite pattern may not be visible in the microstructure, owing to the fact that the primary dendrite austenite is continuous with the eutectic austenite (or their corresponding transformation products exhibit continuity). The same is true of low-carbon irons, but the graphite flakes of necessity outline the dendrites and make their location visible.

We now have a complete statement of the solidification of cast irons yielding flake graphite, according to the stable system. This will provide a defined basis for further discussion. Immediately after solidification, hypo-eutectic and hypereutectic cast irons consist of graphite in a matrix of austenite. Further transformations taking place on cooling will, in practice, depend upon the cooling rate and the silicon content. Two general cases can be mentioned here. If the alloy is cooling slowly, giving stable phases only, or if the silicon content is fairly high (say 3·6%), on cooling below the eutectic temperature carbon is deposited from the austenite in the form of graphite, either upon the already existing graphite or in the form of a Widmanstätten structure. The factors affecting which structure will be obtained have been defined elsewhere [10] (*loc. cit.*, p. 201P). This process will continue until a temperature, the "eutectoid" temperature, is

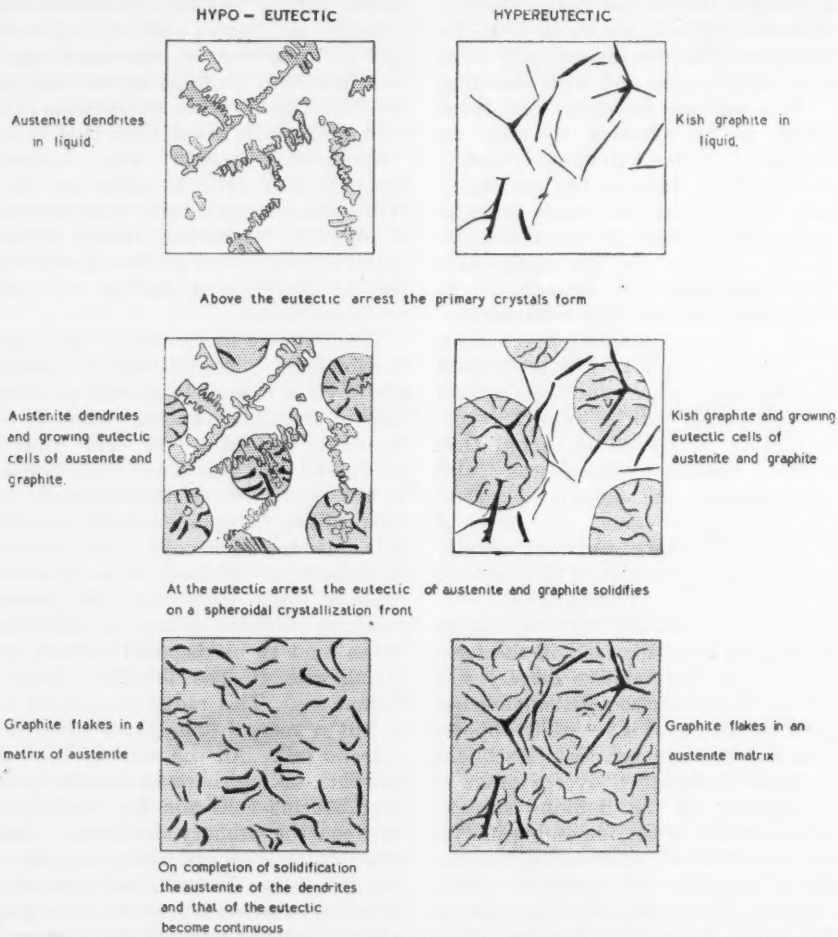
HYPO – EUTECTIC
HYPEREUTECTIC
Austenite dendrites in liquid.
Kish graphite in liquid.
Above the eutectic arrest the primary crystals form
Austenite dendrites and growing eutectic cells of austenite and graphite.
Kish graphite and growing eutectic cells of austenite and graphite
At the eutectic arrest the eutectic of austenite and graphite solidifies on a spheroidal crystallization front
Graphite flakes in a matrix of austenite
Graphite flakes in an austenite matrix
On completion of solidification the austenite of the dendrites and that of the eutectic become continuous

FIG. 3—Progress of solidification according to the stable system in iron–carbon–silicon alloys (or cast irons) with not more than 6% of silicon

reached when the austenite simultaneously deposits α iron (ferrite) and graphite over a range of temperature—the critical range. The range of temperature increases and the temperature limits of this critical range rise with increasing silicon. No graphite eutectoid has definitely been seen and this graphite is usually also deposited upon the already existing graphite flakes. When all transformations occur according to the stable system, the microstructure at atmospheric temperature has graphite in a matrix of ferrite. As a general rule, castings—except those with more than about 4·5% of silicon—do not give stable transformation products, as they are not cooled under equilibrium conditions. Frequently, hypereutectoid graphite is deposited on the already existing graphite flakes, but at the graphite, γ iron (austenite), δ iron (ferrite), and liquid can coexist in equilibrium. On reaching the eutectic concentration an alloy containing, for instance, 7% of silicon will form the δ-iron/graphite eutectic. This will continue until the point of invariant equilibrium is reached, when the melt will react with the δ solid solution to give γ solid solution. When all the δ has been transformed, the melt will continue to deposit the austenite–graphite eutectic. With about 9% of silicon the same cycle of changes occur, but in this case the reaction between the melt and δ will proceed with the complete elimination of the melt, and so no austenite–graphite eutectic is formed. The deposition of graphite probably accompanies the reaction between liquid and δ, and this graphite tends to occur in fine nodules. With

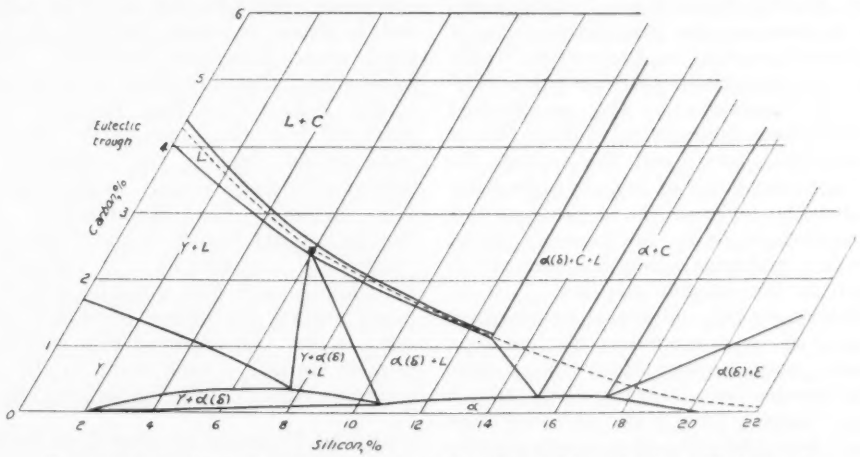

Fig. 4—An isothermal section, just above the plane of invariant equilibrium, of the iron–carbon–silicon diagram. Largely schematic (Jass and Hanemann)

eutectoid temperature range the austenite transforms into pearlite (the cementite–ferrite eutectoid), giving a resulting structure of graphite in a matrix of pearlite. At other times the cooling rate and composition are such that the austenite in the proximity of graphite transforms giving stable products, whilst austenite more distant from the graphite flakes transforms giving metastable products. The net effect of this latter condition is to give a structure of graphite flakes in ferrite lakes in a matrix of pearlite.

The process of solidification described so far is illustrated diagrammatically in Fig. 3. It should be remembered that this is intended to apply only to iron–carbon–silicon alloys (or cast irons) containing less than about 6% of silicon. According to the work of Jass and Hanemann,[16] the point of invariant equilibrium occurs at about 6% of silicon and 2·4% of carbon. At this point 12% of silicon the δ–graphite eutectic only will form. Fig. 4, taken from the work of Jass and Hanemann and incorporating the suggestions of Scheil,[17] shows an isothermal section of the iron-rich corner of the iron–carbon–silicon diagram at about 10° C. above the transition plane of non-variant equilibrium and illustrates the location of the eutectic trough, the γ + α + liquid and the α+ graphite + liquid three-phase regions, and the point of invariant equilibrium.

The Anomalies of Graphite Formation

Undercooled Graphite

Thus far, graphite formation in cast iron during solidification has been dealt with solely from the simple point of view of the formation of flake graphite. However, eutectic graphite occurs in another very common form, variously termed supercooled, eutectiform, reticular, undercooled,

etc. Typical examples of graphite of this type in cast irons are clearly shown in Figs. 21, 25, 29, 33, and 64, where the whole of the graphite is in this form. More frequently, in commercial cast irons, flake and undercooled graphite are both formed. The undercooled form shows the greatest tendency to occur in more rapidly cooled parts, that is, in thin sections and at the cast surfaces. Usually when the two forms occur together the undercooled graphite is positioned at the centres of the eutectic cells, with the coarser flake graphite at their extremities (Fig. 16). It would seem that in these cases the undercooled fine graphite derives from the first eutectic to solidify. This undercooled graphite does occasionally, however, occur in sharp juxtaposition to very coarse flake graphite at the central zones of a casting section, that is, in the most slowly cooled portions (Figs. 17, 18, and 19). Frequently in this case the graphite particles of the undercooled graphite at the centre of the section are finer than those near the edge. The mechanism of formation of the undercooled graphite, the factors determining whether this or flake graphite will form, and the reasons for variations in the tendency of irons of apparently the same chemical composition to show varying degrees of carbide stability, are among some of the outstanding problems relevant to the metallurgy of graphite formation in cast iron.

The literature on this subject is extensive, a large amount of experimental work has been accomplished, many observations on routine commercial practice have been recorded, and as a result many theories have been proposed from time to time to explain the phenomena encountered. It will be generally acknowledged that no theory has yet been successful in explaining all the facts. The whole problem has very many facets and, confronted with this, the present authors feel that investigators have tended to devote their attentions to particular aspects. In what follows an attempt will be made to put together as many as possible of the various anomalies which have to be explained, and to correlate the various theories which have been proposed. In addition, various relevant observations made during routine work in the laboratories of the British Cast Iron Research Association will also be included.

Production of Undercooled Graphite by Superheating and by Titanium and CO_2 Treatment

Fine (undercooled) graphite structures have been produced and recorded by a large number of investigators [23-30], and individual reference to all cannot be made. It was observed and confirmed by many that superheating of the melt favoured the formation of undercooled graphite. Norbury [30] showed undercooled graphite irons to have a greater tendency to chill than flake graphite irons of the same composition. The same investigator also pointed out that the eutectic arrest for fine graphite irons took place at lower temperatures than that for flake graphite irons. Norbury and Morgan [31] showed that an iron normally solidifying in a given section to give coarse flake graphite could be made, by an addition of titanium in a suitable form, to yield fine undercooled graphite. It was found that this process could be considerably enhanced by bubbling carbon dioxide through the melt. The same investigators also found that by bubbling hydrogen through the melt, coarse graphite could be obtained when otherwise fine undercooled graphite would be produced. The titanium could be added to the melt in almost any form, but that giving the best yield was ferrosilicon-titanium. Subsequent experience of these investigators in the laboratories of the British Cast Iron Research Association indicated that the presence of titanium in pig irons would explain the apparently inherent tendency of some brands to give fine graphite. For instance, the Scandinavian pig irons Norskalloy and Vantit, both of which contained appreciable quantities of titanium (and vanadium) gave fine graphite structures after many melting operations, including cupola melting. It was also observed that bubbling carbon dioxide through a melt, even in the absence of titanium, will tend to refine the graphite structure.

Forms of Titanium in Cast Iron, and Effect of Zirconium

In irons of normal nitrogen content and containing sufficient manganese to balance the sulphur content, titanium exists in the microstructure as titanium carbide.[32] One of the present investigators (H. M.) has observed that unless the titanium is present as carbide, no refining action is given. When the titanium is fixed as titanium sulphide (owing to insufficient manganese to balance the sulphur), and probably also when completely fixed as titanium cyanonitride, then coarse graphite structures will be obtained. This fact is significant in view of a further finding of the present investigators, *viz.*, that zirconium behaves in a manner similar to that of titanium. In view of the chemical similarity of these two elements, this fact is not surprising. It has been shown [32] that the function of zirconium when added to a normal grey cast iron is to form a probably complex sulphide phase. When the zirconium content exceeds about 0·1–0·2%, the zirconium carbide, ZiC, appears,

and zirconium only seems to exert a positive refining effect when this carbide phase is present. Figs. 20 and 21 illustrate the graphite-refining effect of zirconium. Fig. 20 shows the coarse flake graphite structure in a 3-in. dia. bar, untreated and having the following analysis :

T.C.	Si	Mn	S	P
3·09%	1·58%	0·93%	0·018%	0·05%

Fig. 21 shows the undercooled graphite structure in a 3-in. dia. bar after treatment with ferrosilicon-zirconium to give the analysis :

T.C.	Si	Mn	S	P	Zr
3·13%	3·05%	0·81%	0·011%	0·054%	0·55%

It seems, therefore, to be definitely established that the refining effect of both of these elements is connected with the formation of their carbides.

Von Keil's Silicate Slime, and Norbury and Morgan's Inclusion Theory

Von Keil and his co-workers [33] have attributed the transmission of anomalous characteristics from the melting stock to the casting to the presence of iron silicate in the pig iron, and have suggested that a submicroscopic silicate slime nucleates the melt to give coarse flake graphite, whilst the absence of these fine particles allows of the formation of fine undercooled graphite. Norbury and Morgan,[31] in order to explain the effect of titanium and carbon dioxide, suggested that this process modified the chemical nature of the " inclusions " (meaning silicates and oxides, these being similar to von Keil's silicate slime) in such a manner that they became liquid, whereas they would normally be solid. In this state they could not nucleate the melt, and so fine graphite was produced. Further to support this argument these investigators cited the refining action of copper in excess of liquid solubility. In this case free molten copper was present during the solidification of the eutectic, and so permitted the formation of fine undercooled graphite by surrounding the solid inclusions.

Following this work the literature became liberally interspersed with allusions to " slag inclusions " and " silicate slimes " without any real attempt being made to define these particles with any precision. The idea received some support in the work of Diepschlag [34], who found that melting in contact with aluminous slags had a graphite-refining effect and melting in contact with siliceous slags a coarsening effect. Norbury and Morgan (*loc. cit.*) were unable to confirm this, but it is undoubtedly the experience of the ironfounding industry that slag composition does influence the chilling characteristics and the graphite size of cast iron. The validity of the silicate-slime theory has yet to be proved or disproved, but it will undoubtedly require considerable modification in order to accommodate the other relevant facts. Considerable difficulty has been experienced in advancing this theory further, owing to the difficulty of establishing the identity of the particles necessary for the theory. Microscopically, cast iron is characterized by the almost complete absence of visible oxides and silicates under normal conditions.[32] It may seem surprising that in this respect cast iron and pig iron should be so very much cleaner than steels, but a study of slag–metal equilibria indicates that this should be so. The immiscibility of slag and metal increases at lower temperatures ; at lower temperatures the probability of a slag being solid is greater. The effect of carbon is to increase the region of immiscibility in slag–metal systems. Cast iron tends to be melted and cast at much lower temperatures than steel and in addition has a much higher carbon content, and so the separation of slag is facilitated. By superheating cast-iron melts to over 1700° C., followed by fairly rapid cooling, it has recently been found possible by the present investigators to retain some small particles of glassy silicates in the microstructure. Chemical analysis undoubtedly indicates the presence of oxygen in cast iron, and considerable interest attaches to the mode of occurrence of this element.

In the course of a very lengthy investigation carried out by Norbury and Morgan at the British Cast Iron Research Association, it became apparent that the production of fine graphite by the titanium/carbon-dioxide process* had important influences on the properties of some irons. In irons of coarse graphite and with a relatively high carbon content, an improvement in mechanical properties could be obtained by the refining treatment. Still more important, it was discovered that the production of fine graphite by this method gave sound castings which would otherwise have been porous.[35] This effect is very pronounced, and no doubt explains the soundness of the castings made in permanent moulds by the Eaton–Erb or Holley process.[36] The very rapid cooling of these castings invariably gives fine undercooled graphite. Whilst irons refined by the titanium/carbon-dioxide process are relatively very free from porosity, they show an enhanced tendency to sink on the top surfaces of the casting. Porosity and sinking occur during the solidification of cast iron, and since these vary in amount and type according to whether the graphite is of the flake or undercooled type it seems logical to expect that the mechanism of solidification is

* British Patent Specification 425,227.

different in the two cases. This will be dealt with in greater detail later.

Inoculation

At the same time that attention was focused on the factors affecting the production of undercooled graphite, considerable progress was being made in the production of high-duty irons by the use of controlled low carbon contents of between 2·7 and 3·3%. This experience indicated that undercooled fine graphite structures were unsatisfactory, and led to the development of processes the aim of which was the elimination of these structures and their replacement by flake graphite. This was achieved by a procedure now referred to as "inoculation," which involved treatment of the melt with a relatively small amount of a substance, generally a graphitizer, immediately before pouring. The characteristic of these inoculants is the fact that, added in this way, they have a graphitizing effect much greater than when included in the charge with the other melting stock. This reduction in chilling tendency is accompanied by the elimination of fine undercooled graphite for normal cooling rates. It is impossible to indicate who originated this process, but the use of calcium silicide[37] (which became the basis of the Meehanite process) and ferrosilicon were among the first methods to be adopted. Others have subsequently been developed, such as the use of S.M.Z.,* silicon carbide, and graphite. Almost without exception the graphitizing inoculants contain either silicon or carbon. A feature of the inoculating process is the fact that the effect wears off and does not persist on remelting. If the metal be held in the molten condition too long before casting, normal chilling and undercooled graphite will result.

Undercooling Theory

It has been demonstrated by many investigators[30, 40–43] that eutectic arrest in irons yielding undercooled graphite takes place at a lower temperature than in irons yielding flake graphite. Cooling curves for irons with mixed structures show a recalescence after the initial arrest. These facts, together with the slag-cloud theory of von Keil and that of Norbury and Morgan, to mention only two of the most influential research schools, caused the gradual evolution of a theory which argued that the fine graphite structures were produced by undercooling of the melt in the absence of sufficient or appropriate nuclei.

It is appropriate to consider here the background on which the undercooling theory was evolved. It is well known that pure liquid metals will readily undercool to temperatures below the normal freezing point until a point is reached when crystallization suddenly begins and the temperature rises. This effect can be minimized, and even avoided, by agitation or by nucleating the melt with solid particles which may or may not be of the same composition as the melt. The effect of undercooling† on the form of a cooling curve is illustrated in Fig. 5. In some organic and inorganic compounds the undercooled state may be preserved indefinitely at atmospheric temperatures to give glassy, isotropic, amorphous "solids." No case is known of this in intermetallic alloys. It has been shown for solutions that the region below the normal solubility curve may be divided into two zones. In the region immediately below this curve crystallization will not occur spontaneously, but requires nucleation of the melt. At a certain fixed temperature below the true solubility curve, a zone is reached where crystallization begins without nucleation. For mixtures it is possible to draw a curve, delineating the zone of spontaneous crystallization, which roughly lies parallel to the normal solubility

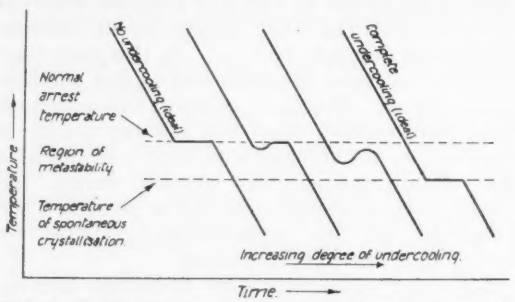

FIG. 5—Effect of undercooling on the form of the cooling curve

curve and which is termed the supersolubility curve. Our knowledge of undercooling in simple eutectiferous systems is limited to the results of Miers and Isaac[44, 45] for the two organic substances betol and salol, and Desch[46] has suggested

* S.M.Z. is a proprietary silicon–manganese–zirconium alloy[39] containing approximately 63% of silicon, 6% of manganese, 6% of zirconium, and 20% of iron. The precise function of the deoxidizing elements is obscure, but this alloy gives better results than ferrosilicon used in amounts to give similar yields of silicon. Significantly however, the performance of ferrosilicon can be improved by mixing with aluminium turnings,[39] and one of the present authors finds this even better than S.M.Z.

† Frequently the term "supercooling" is used synonymously with "undercooling," but the latter is preferred by the present authors as it is felt that "supercooling" carries an implicit contradiction. It is realized that many preferring "supercooling" relate it to supersaturation, supersolubility, and superheating.

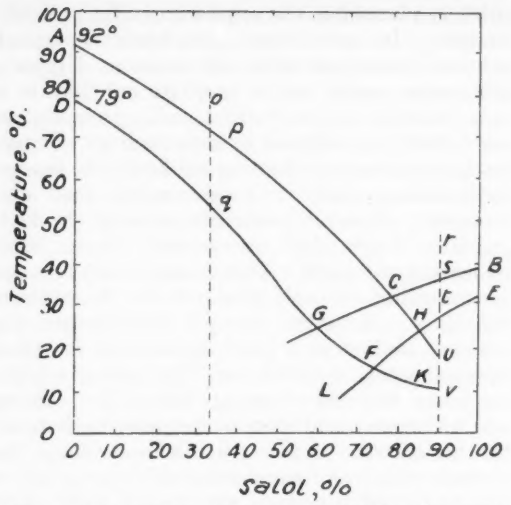

FIG. 6—The system betol–salol, showing the supersolubility curve (Desch)

that these conditions may also apply to metallic alloys. The diagram giving the normal and supersolubility curve for betol and salol (Fig. 6) is reproduced from Desch. The normal eutectic point is at *C*. The two supersolubility curves *DF* and *EF* intersect at a point *F*, termed the hypertectic point. The eutectic and hypertectic need not have the same composition. The area *ACBEFD* is the metastable region requiring nucleation to initiate crystallization. Below *DFE* is the region of spontaneous or labile crystallization. The following quotation from Desch indicates the several possible ways which solidification may proceed owing to undercooling in such a system :

"A mixture represented by the point *o* is cooled, the access of solid matter being prevented. When the point *p* is reached, the solid phase being absent, crystallization does not take place, and the mixture remains liquid until the temperature has fallen to *q*. The supersolubility curve of betol, or more generally of the substance *M*, has now been cut, and crystallization takes place spontaneously if the mixture is shaken. The composition of the part remaining liquid is now represented by a point travelling down *DF*, until *F*, the hypertectic point, is reached. This being the point of intersection with the second supersolubility curve, the second solid, *N*, begins to separate and the whole of the mother-liquor solidifies at the constant temperature of the point *F*.

Taking the same mixture *o*, we may now assume that on reaching the temperature *p*, a crystal of *M* is introduced. This crystal grows, and the point representing the composition of the liquid moves along *pC*. When the eutectic point *C* is reached, however, the eutectic mixture does not solidify as a whole, as it is metastable as regards the substance *N*, the separation of *M* therefore continues in the direction *CH*. The temperature then falls until the curve *EF* is cut, crystallization of *N* begins, and continues until the hypertectic point is reached. The curve *BC* may similarly be prolonged into the metastable region, as shown by *CG*.

In the absence of mechanical disturbance, mixtures may even be cooled below the temperatures indicated by the curves *DF* and *EF* without crystallizing. Thus the mixture *o* may be cooled, under favourable conditions, below *q*, but crystallization is sure to begin before the undercooling has proceeded very far into the labile region, unless prevented by increasing viscosity. The heat developed causes the temperature to rise to *q*, and freezing then proceeds. Under similar conditions, however, the phase *N* may also fail to appear when *F* is reached, so that the labile prolongation *FK* may be experimentally realized. The curve *EF* may be similarly prolonged as at *FL*.

The formation of crystals on the upper curve takes place with comparative slowness, and the temperature may therefore fall low enough for the supersolubility curve to be intersected before the separation of solid is complete. In that case a further crystallization of the same solid phase, but now proceeding rapidly from a large number of centres, may take place. Miers and Isaac found it easy to distinguish by inspection alone the crystals formed in the metastable region from the shower deposited on reaching the supersolubility curve.

It is clear from this description that the eutectic point *C* may entirely fail to appear on the cooling curve of a mixture. In one case, that of a mixture containing 90% of salol and 10% of betol, Miers and Isaac were even able to observe four freezing points, none of which was the eutectic point. Representing such a mixture by *r*, the upper salol curve is crossed at *s*, at which point salol crystals may be obtained by inoculation. Crystals of betol are obtained at *u* by inoculation with that compound, whilst labile showers of the two components may be obtained spontaneously by cooling to *t* and *v* respectively."

It should be noted here that these are not all theoretical possibilities and, as will be shown later, they may be pertinent to the solidification of cast irons.

Possible Modes of Undercooling in Cast Irons

It will be seen that the fine graphite structures obtained by superheating, by rapid cooling, or by the titanium/carbon-dioxide process could perhaps be explained by undercooling according to the pattern outlined above. In the absence of nuclei it may be assumed that hypo-eutectic cast irons could undercool to the supersolubility curve, austenite being deposited down the supersolubility curve until the hypertectic point is reached and its corresponding binary complex is formed (Fig. 7*a*). If austenite formed along the normal liquidus and continued along its extension until the supersolubility curve on the hypereutectic side was reached, another phase would have to be deposited before the hypertectic point was reached (Fig. 7*b*). The present authors

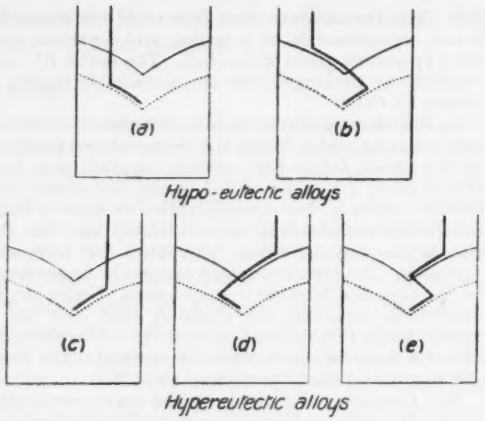

FIG. 7—Possible ways by which undercooling could occur in cast irons

have no definite knowledge of structures indicating that this has occurred, although when taking cooling curves of 50-g. samples of iron giving fine graphite, a change in slope was observed in some cases on the portion of the curve immediately before the eutectic arrest. A typical case is shown in Fig. 8. This change of slope begins at approximately the temperature of the normal eutectic, and may indicate the deposition of supersaturated austenite and/or a phase from the hypereutectic portion of the supersolubility curve.

Three similar cases of undercooling could apply to hypereutectic alloys (Fig. 7, *c*, *d*, and *e*). Diagrams *a* and *b*, for hypo-eutectic alloys, are essentially similar to diagrams *c* and *d*, for hypereutectic alloys. A case, similar to diagram *e*, can be constructed for hypo-eutectic alloys. Some metallographic evidence is available that the type may occur involving undercooling of hypereutectic alloys along the extension of the line indicating the solubility of graphite in the melt until the supersolubility curve for austenite is reached, and then down the supersolubility curve to the hypertectic point (Fig. 7*c*). This type of undercooling would involve primary phases being deposited characteristic of both sides of the eutectic. Some hypereutectic cast irons do show kish graphite and transformed primary austenite dendrites. Nipper[8, 9] has shown some excellent micrographs illustrating this effect with more emphasis than in any sample prepared by the present authors, and Fig. 22, which is a reproduction of one of Nipper's micrographs, illustrates the point. A large flake of primary kish graphite is seen crossing the field, and this flake is completely insulated from the undercooled graphite by a layer of graphite-free metal which in places has the appearance of a primary dendrite. In some cases completely separated kish and primary dendrites can be seen. Nipper's melts were carried out in graphite crucibles in a high vacuum (below 0·0001 mm. pressure) at about 1640° C., followed by slow cooling. Superheating and vacuum heating are known to favour undercooling, and it would seem that the deposition of metal (austenite) around the kish graphite flakes had prevented them from nucleating the melt. This is not always so, as the following example prepared by the authors will show. A 50-gm. sample of hematite pig iron was melted in a graphite crucible *in vacuo* (approximately 0·00001 mm. Hg), using a high-frequency induction-heating unit. The cooling rate was more rapid than that quoted by Nipper. At the bottom of the crucible and along the vertical walls, well-formed flakes of kish graphite—apparently radiating from the crucible walls—had inoculated the melt, and no undercooled graphite was present (Fig. 23). In the upper part of the melt a segregation of broken kish graphite flakes occurred (Fig. 24), owing to flotation from the lower and central zones. The graphite had presumably been mutilated by turbulence of the melt during flotation. Again, no undercooled graphite was present. The central zones from which the kish graphite had been removed by flotation showed considerable areas of undercooled graphite with well-defined transformed primary austenite dendrites (Fig. 25). Except for one feature (which will be dealt with later), this structure appears typically hypo-eutectic. This melt was saturated with carbon at a high temperature, and, except on the basis of undercooling, it is difficult to see how one portion of the structure could appear hypo-eutectic.

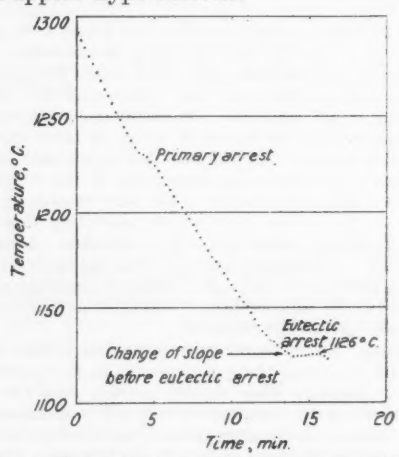

FIG. 8—Inflection on cooling curve immediately before the eutectic arrest

The deposition of a layer of austenite around the primary hypereutectic graphite can be explained by a process other than that referred to above. It is well known that when large and small crystals are in contact with a melt there is a tendency for the larger to continue to grow at the expense of the smaller. This is particularly so when we have primary dendrites surrounded by a solidifying eutectic, the dispersed phase of which is the same as the dendrites. In these cases, in the immediate vicinity of the primary dendrites, the corresponding phase of the eutectic is deposited upon the already existing dendrite, causing it to be surrounded by an unbroken layer of the continuous phase and therefore apparently separated from the eutectic. Desch[48] has illustrated this feature in copper–phosphorus alloys, but iron–carbon alloys are especially interesting in this respect, as is indicated in what follows. In the white irons the dispersed phase of the eutectic is austenite and the continuous phase is cementite, so that white irons of hypo-eutectic composition show this phenomenon. In grey irons the dispersed phase of the eutectic is graphite and the continuous phase is austenite, so that only hypereutectic grey irons would be expected to show this effect. At the eutectoid temperature in white irons the dispersed phase of the eutectoid is cementite and the continuous phase is ferrite, and, as a result, the eutectoid cementite frequently crystallizes on to the already existing eutectic cementite, giving the apparent anomaly of ferrite adjacent to cementite.

From what has been said so far on this point it will be apparent that it is not clear whether undercooling causes the formation of austenite around the graphite, or whether austenite forming round the graphite permits undercooling by insulating it from the melt.

At this stage it should be noted that later in this paper undercooled graphite will be demonstrated to arise from the decomposition of the carbide of a carbide–austenite binary complex shortly after solidification. If the scheme of undercooling outlined above is correct then the hypertectic point is that at which this carbide–austenite binary complex solidifies. Furthermore, the phase deposited down the hypereutectic side of the supersolubility curve must also be a carbide. The unique implications of this will be dealt with more fully in a later section.

Graphite-Nuclei Theory

The silicate-slime theory of von Keil and that of Norbury and Morgan could not explain the inoculating effects of ferrosilicon, calcium silicide, and graphite. Piwowarsky[23] developed a graphite-nuclei theory to explain the increased tendency to undercooling as the result of superheating. This process caused the solution of these graphite particles. This theory was considerably weakened when the same investigator presented evidence[49] to show that graphite dissolves extremely rapidly above the liquidus. Recently, however, the theory has received further support, by Massari and Lindsay,[50] who claim to have demonstrated the presence of traces of residual free carbon in remelted material which decreased in amount with increase in temperature of superheat. Metal from the bath was dribbled directly into cold water and the resulting shot was sectioned for metallographic examination. In support of their claims these investigators produced micrographs which showed black spots, which were termed graphite nuclei. No special tests, such as the determination of the optical properties under polarized light, were carried out to identify this phase as graphite. The present authors emphasize the extreme value of such tests in support of claims which might be put to doubt. Massari and Lindsay also confirmed the powerful inoculating (chill-removing) effects of graphite additions to the melt. This latter point is now well established in commercial practice. Eash[43] has also added his support to the graphite-nuclei theory by suggesting that the inoculating effect of ferrosilicon is due to localized silicon concentrations reducing the carbon solubility. A localized silicon concentration might cause the iron to become hypereutectic and so to deposit kish graphite, which nucleates the melt. It is conceivable that calcium silicide might function in this manner, but it is difficult to maintain that treatment with hydrogen, which also produces flake graphite, would cause local deposition of graphite. This effect is more susceptible to treatment by Norbury and Morgan's theory.

Effect of Gas and Sulphur on Formation of Graphite—The Work of Boyles

In a remarkable series of papers, Boyles[20, 42, 51, 52] demonstrated the mechanism of solidification of hypo-eutectic cast irons. The description already given in this paper is largely in agreement with the findings of this investigator. He demonstrated that hypo-eutectic iron–carbon–silicon alloys and cast irons when melted *in vacuo* tend to give undercooled graphite structures. When remelted in hydrogen this vacuum-melted material was found to give coarse flake graphite, but iron–carbon–silicon alloys behaved somewhat differently from cast irons. This investigator also

indicated that the coarsening effect produced by remelting undercooled graphite alloys in air or nitrogen is due to the absorption of hydrogen, as water vapour, from these gases.

Another very important feature revealed by Boyles was that iron–carbon–silicon alloys which normally give the modified structure behave as cast irons giving coarse flake graphite when an addition of sulphur, or sulphur and manganese, is made. This is a most important finding, and Boyles has suggested that the effect of sulphur is related to its carbide stabilizing action. With sulphur additions beyond a certain percentage it was found that the coarsening action ceased. With reference to the findings of Norbury and Morgan, Boyles suggested that the effect of carbon dioxide in giving undercooled structures was due to its scavenging action in removing hydrogen from the melt. The effect of sulphur additions in causing the formation of large amounts of manganese sulphide in the melt, and so promoting the formation of coarse flake graphite, has been confirmed by the present authors.

That manganese sulphide may nucleate a melt seems highly probable, and the present authors have observed manganese sulphide segregations associated with patches of coarse graphite. Unfortunately, Boyles did not give any precise indications of the types of sulphide phases present in all of his melts. In one series of melts various additions of sulphur were made to base material which contained no manganese. In this case it seems logical to assume iron sulphide to be present. Under these conditions, increasing sulphur gave progressively coarser graphite sizes until a maximum was reached beyond which the graphite became finer. Iron sulphide containing no manganese is liquid during the eutectic solidification, and therefore the coarsening effect cannot be due to solid nucleation. In cast irons with high carbon and silicon contents it is possible that the melt is not completely miscible with liquid iron sulphide, so that some free sulphide may exist. Oxides and silicates, particularly FeO, are known to be soluble in iron sulphide, so that this compound may remove oxygen from the metal by partition.

Eutectic Cell Size and Nucleation

Boyles has also demonstrated that the eutectic cell size is greater and that these cells are much less numerous in undercooled irons than in flake graphite irons. To a considerable extent this fits the facts relevant to the nuclear theory. The number of eutectic cells in a given sample precisely indicated the number of centres at which eutectic crystallization begins. If undercooling in cast iron occurs owing to the absence or deficiency of nuclei, then fewer eutectic cells are to be expected. It is frequently wrongly supposed that nucleation of cast-iron melts solely implies the initiation of the crystallization of graphite. However, in this case we are really concerned with initiating the crystallization of a eutectic. That the graphite phase of the eutectic is the important one for the nucleation is highly probable, and no case has come to the authors' knowledge of austenite initiating the crystallization of the normal flake graphite eutectic. The number of graphite flakes present in a sample is not primarily or directly a function of the state of nucleation of the melt. Nucleation of the melt merely affects the number of centres at which eutectic crystallization begins, and as this determines the size of the eutectic cells it has an indirect effect on graphite size. The radius of a eutectic cell represents the maximum possible length of a eutectic graphite flake, and as the carbon content of the alloy is reduced the realization of this maximum is reduced, owing to a higher probability of an austenite dendrite opposing eutectic crystallization in any one direction. When the nucleation and carbon content of a melt are fixed, the graphite-flake size then becomes a function, within the above limits, of the cooling rate and (an associated factor) turbulence of the solidifying melt. It has frequently been wrongly supposed that the fine undercooled graphite flakes being present in large numbers indicates a high state of nucleation, but the number of eutectic cells indicates the state of nucleation and the uniform fine nature of the graphite is an indication of the spontaneity of its formation. It should be noted that no explanation of the formation of this fine graphite has been given, except in so far as undercooling is a necessary concomitant.

In what has been said so far the term "nucleus," implied by "nucleate," has not been used in any other sense than that indicating a location in the melt where physical and/or chemical conditions are most favourable to crystallization. The value of a particular location for initiating crystallization must depend upon the cooling rate, that is, upon the extent to which the melt is undercooled below the true liquidus. For inoculated irons there is a considerable range of cooling rates within which the nuclei function to initiate flake graphite crystallization, but with very rapid cooling rates low-silicon irons will become white and high-silicon irons will give undercooled graphite. For instance, in permanent-mould irons both kish graphite and late additions of ferrosilicon or

FIG. 9—Graphite flake in blast-furnace bear, in position of minimum pleochroic absorption. Under plane-polarized light. Unetched × 200

FIG. 10—Same spot as in Fig. 9, but between crossed nicols and with graphite in brightest position, *i.e.*, at 45° to plane of polarization of incident light. × 200

FIG. 11—Same spot as shown in Figs. 9 and 10, but with graphite flake parallel to plane of polarization of incident light, *i.e.*, in the same position as for minimum pleochroic absorption. Between crossed nicols × 200

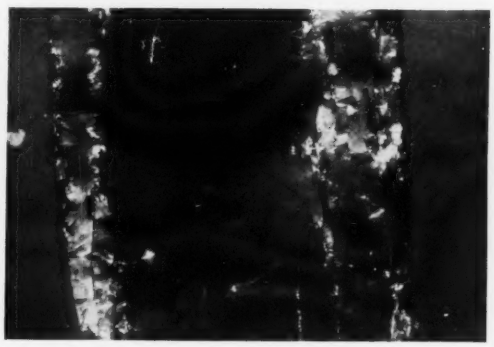

FIG. 12—Same spot as shown in Figs. 9 to 11, but with graphite flake at right angles to the plane of polarization of incident light. Between crossed nicols × 200

FIG. 13—Spot marked off in Fig. 9 under same conditions, but at a higher magnification. × 500

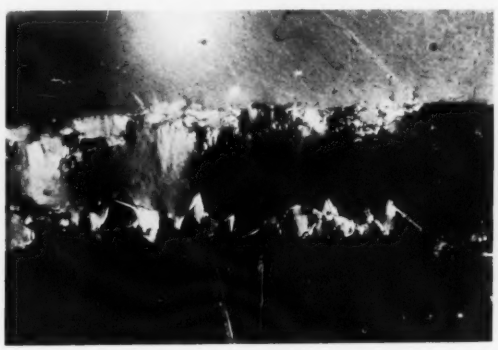

FIG. 14—Spot marked off in Fig. 9 at same magnification as in Fig. 13, but between crossed nicols. × 500

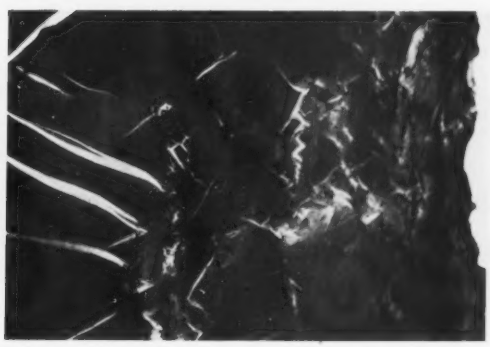

FIG. 15—Flake of kish graphite in blast-furnace bear, with zigzag markings in outer banded layers and at the centre. In position of maximum pleochroic absorption under plane-polarized light. Unetched × 500

(Micrographs reduced to three-quarters linear in reproduction)

[*Morrogh and Williams*
[*To face p.* 336

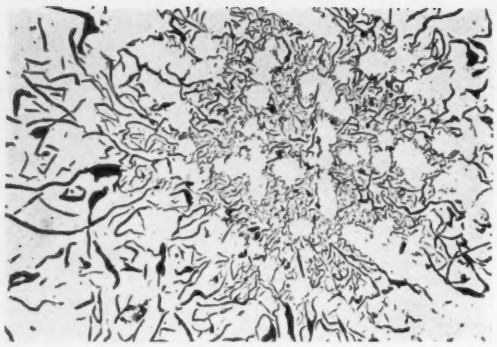

FIG. 16—Cluster-type graphite with undercooled graphite at the centre and normal flake graphite at the periphery of the cluster. Unetched × 60

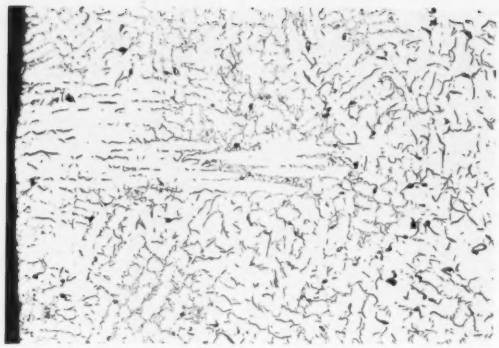

FIG. 17—Graphite structure near the edge of a 1·2-in. test-bar. Fine flake graphite, with traces of undercooled graphite near the edge. Unetched ×60

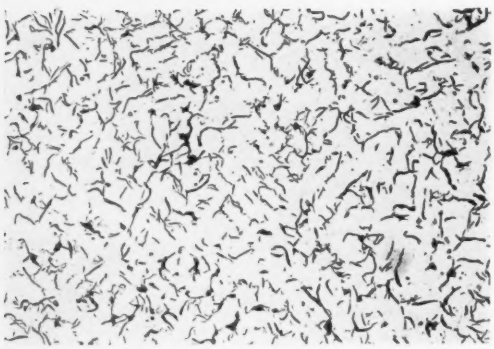

FIG. 18—Same bar as illustrated in Fig. 17, but showing the graphite structure midway between the edge and centre. Medium-sized flake graphite. Unetched × 60

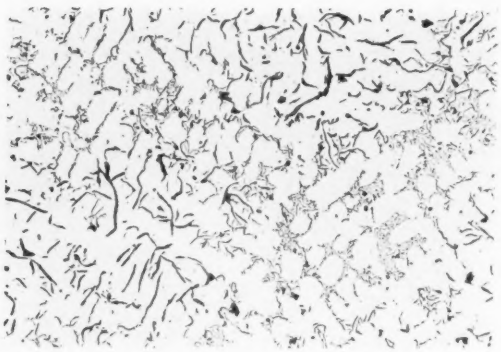

FIG. 19—Same bar as illustrated in Figs. 17 and 18, but showing the graphite structure at the centre of the section. Coarser flake graphite with patches of fine undercooled graphite. Unetched ×60

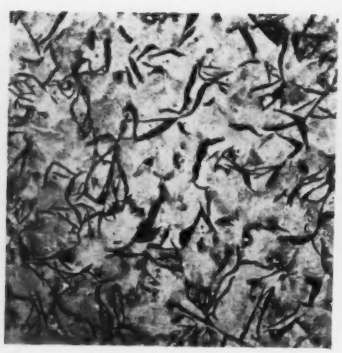

FIG. 20—General structure of 3-in. dia. bar. Untreated. Etched with picric acid × 100

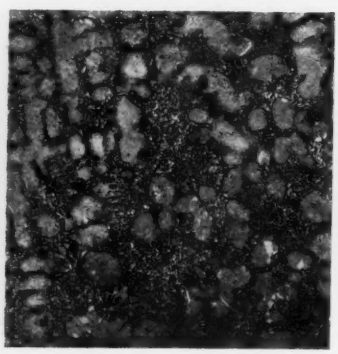

FIG. 21—General structure of 3-in. dia. bar. Same melt as in Fig. 20, but after treatment with ferrosilicon-zirconium. Etched with picric acid × 100

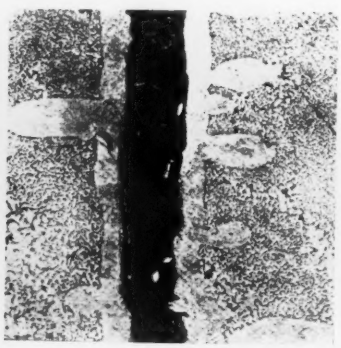

FIG. 22—Primary graphite flake surrounded by transformed austenite and fine eutectic graphite (Nipper). Etched ×200

(Micrographs reduced to three-quarters linear in reproduction)

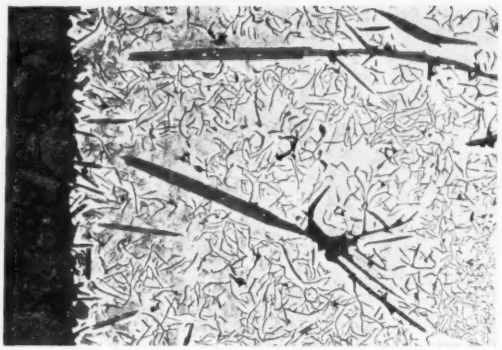

FIG. 23—Well-formed hypereutectic graphite flakes at the portions of ingot near to crucible walls. Ingot was melted in vacuum, using a graphite crucible. Etched with picric acid × 60

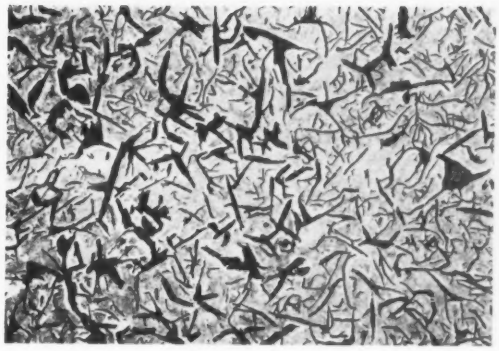

FIG. 24—Coarse graphite and fragmented hypereutectic graphite near the top of vacuum-melted ingot. The kish graphite has inoculated the melt in this zone. Etched with picric acid × 60

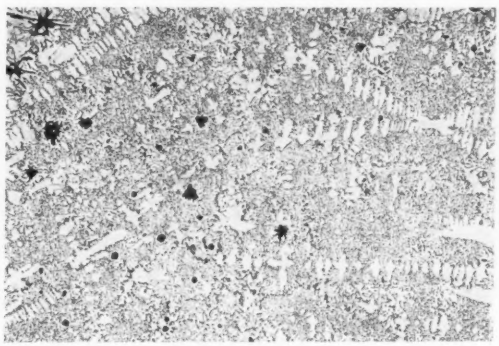

FIG. 25—Central zone of vacuum-melted ingot. Undercooled graphite with transformed primary austenite dendrites. The hypereutectic graphite has floated away from this zone to the top of the ingot. Etched with picric acid × 60

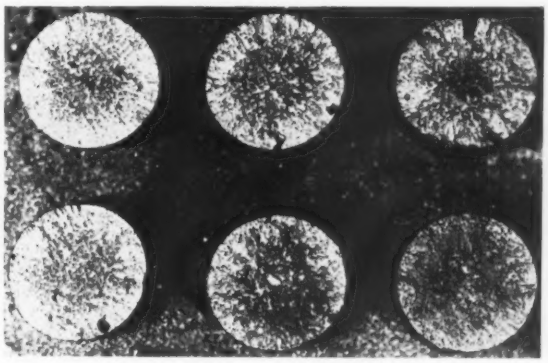

FIG. 26—Bars treated with sulphur, showing effect of pouring temperature on chilling

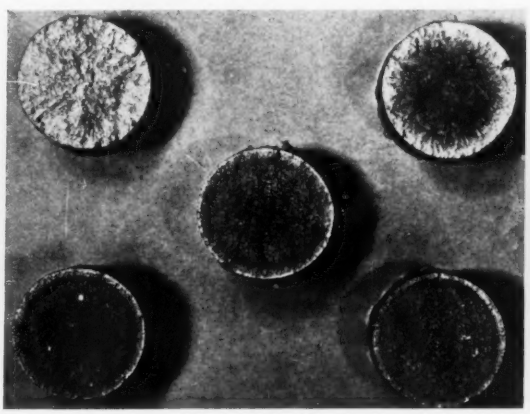

FIG. 27—Effect of pouring temperature upon the chilling of an iron treated with tellurium

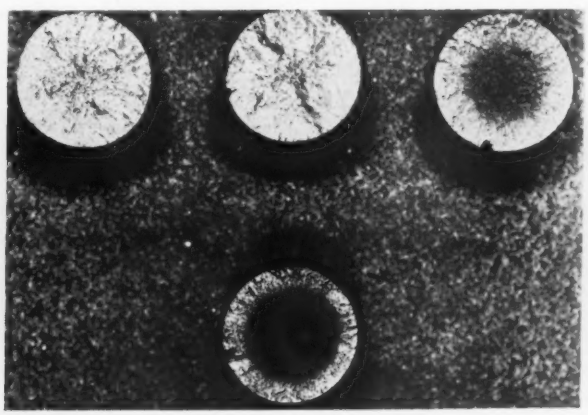

FIG. 28—Effect of pouring temperature upon the chilling of a tellurium-treated iron after remelting

(Figs. 23 to 25 reduced to three-quarters linear in reproduction)

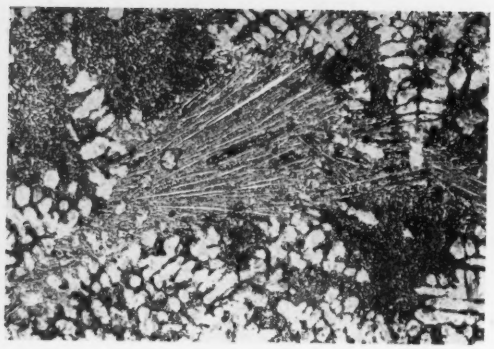

FIG. 29—Inverse chill in an iron with undercooled graphite. Remainder of structure is undercooled graphite in a matrix of pearlite. Etched with picric acid × 60

FIG. 30—Fractures of 0·875-in. bars, showing inverse chill

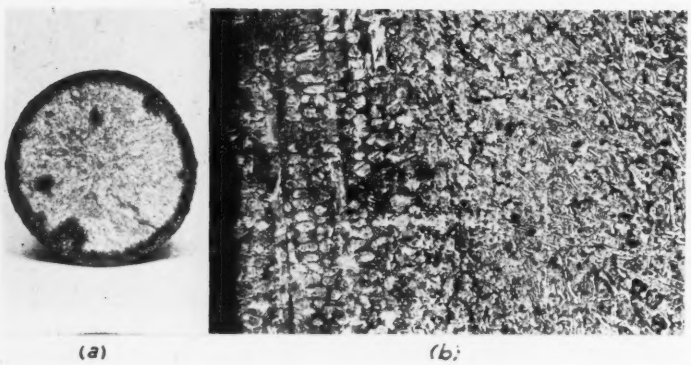

FIG. 31—Influence of sand mould in causing a white iron to have an outer grey rim: (*a*) Fracture of a 0·875-in. dia. test-bar. (*b*) Micrograph of a similar bar showing graphite in the outer rim. Etched with picric acid × 60

FIG. 32—Mixture of flake and undercooled graphite. Unetched × 60

FIG. 33—Same melt as in Fig. 32, but after treatment with 0·2% of bismuth. Entirely very fine undercooled graphite. Unetched × 60

FIG. 34—Hypo-eutectic nickel-carbon alloy. Ingot 1*A*. Melt superheated to 1600° C. Unetched × 60

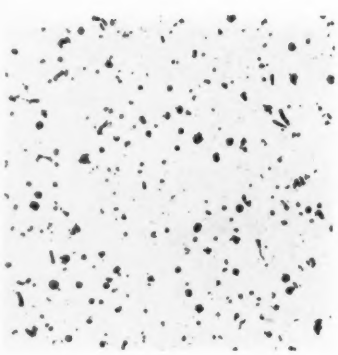

FIG. 35—Hypo-eutectic nickel-carbon alloy. Ingot 1*B*. Same melt as in Fig. 34, but after treatment with calcium silicide. Nodular graphite. Unetched × 60

(Micrographs reduced to three-quarters linear in reproduction)

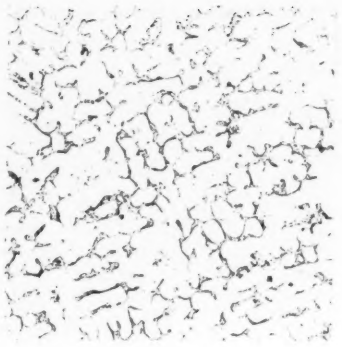

FIG. 36—Hypo-eutectic nickel carbon alloy. Same melt as in Fig. 35, but after treatment with ferrosilicon-zirconium. Unetched × 60

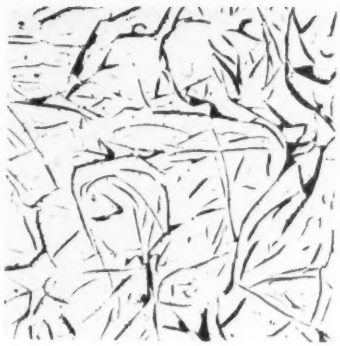

FIG. 37—Hypo-eutectic nickel carbon alloy. Ingot 2*B*. After treatment with tellurium. Coarse flake graphite. Unetched × 60

FIG. 38—Coarse graphite structure in hypo-eutectic nickel carbon alloy after treatment with sulphur. Ingot 4*B*. Small areas of Ni_3S_2 can be seen. Unetched × 60

FIG. 39—Flake graphite structure in ingot 4*C*. Same melt as in Fig. 38, but after addition of 1% of manganese. Unetched × 60

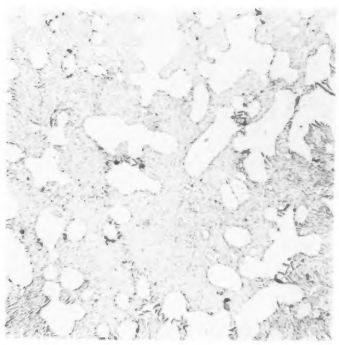

FIG. 40—Hypo-eutectic nickel carbon alloy. Ingot 5*A*. Superheated to 1600° C. Undercooled-type and coarser graphite. Unetched × 60

FIG. 41—Ingot 5*B*. From same melt as in Fig. 40, but after bubbling hydrogen through the melt. Unetched × 60

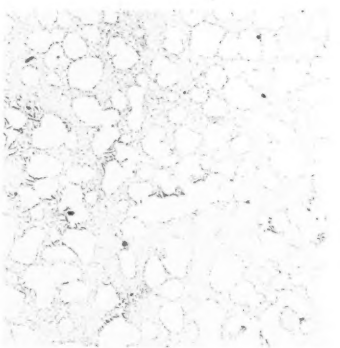

FIG. 42—Undercooled graphite structure in a hypo-eutectic nickel carbon alloy after treatment with calcium silicide and then with hydrogen. Ingot 6*C*. Unetched × 60

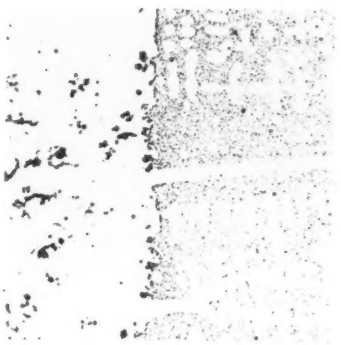

FIG. 43—Undercooled graphite rim at the periphery of ingot 6*B*, treated with calcium silicide. Unetched × 60

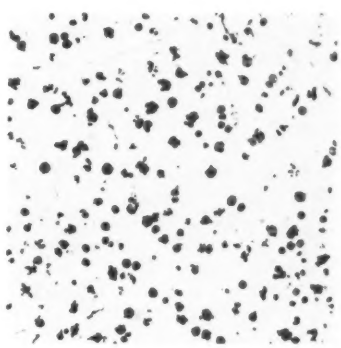

FIG. 44—Hypereutectic nickel carbon alloy. Ingot 8*B*. Same melt as in Fig. 45, but after treatment with calcium silicide. Unetched × 60

(Micrographs reduced to three-quarters linear in reproduction)

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FIG. 45—Hypereutectic nickel–carbon alloy. Ingot 8.4. Superheated to 1600°C. Kish graphite flakes in undercooled graphite eutectic with primary solid-solution dendrites. Unetched × 60

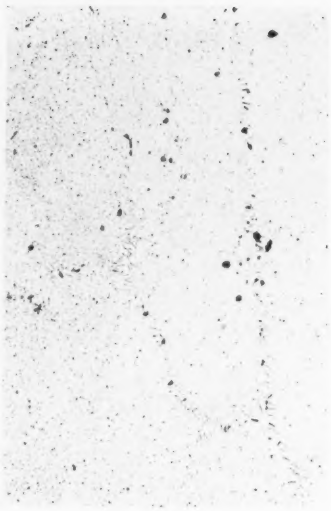

FIG. 46—Wholly eutectic area in ingot 9.4. Hypereutectic graphite has floated to the top of the ingot. Unetched × 60

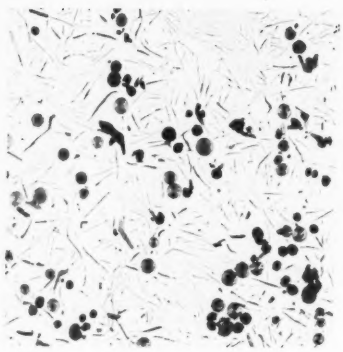

FIG. 47—Area near the top of ingot 9.4; flotation of hypereutectic graphite has occurred. Well-formed spherulites are also present. Unetched × 60

FIG. 48—Hypereutectic nickel–carbon alloy treated with tellurium. Ingot 9*B*. Very coarse flake graphite. Unetched × 60

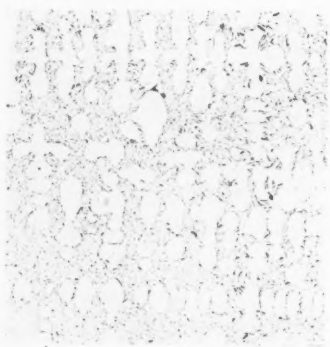

FIG. 49—Hypo-eutectic nickel–carbon alloy. Ingot 10.4, superheated to 1600°C. Unetched × 60

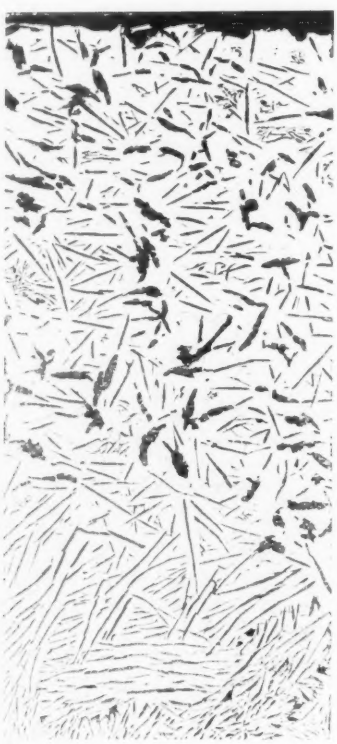

FIG. 52—Segregation of kish graphite at the top of ingot 16. Note absence of undercooled graphite. Unetched × 60

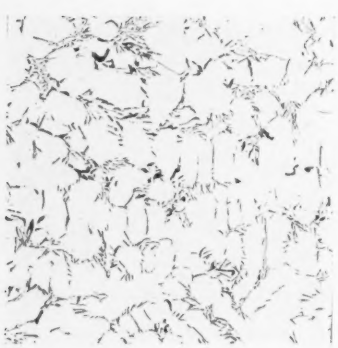

FIG. 50—Hypo-eutectic nickel carbon alloy. Same as in Fig. 49, but after treatment with titanium and CO_2. Unetched × 60

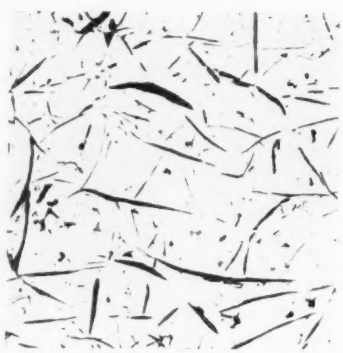

FIG. 51—Ingot made from melt of pure nickel + 2% of C, 2·5% of Si, 1·25% of Mn, 1·3% of P, and 0·3% of S. Coarse flake graphite and lakes of nickel phosphide. Unetched × 60

(Micrographs reduced to three-quarters linear in reproduction)

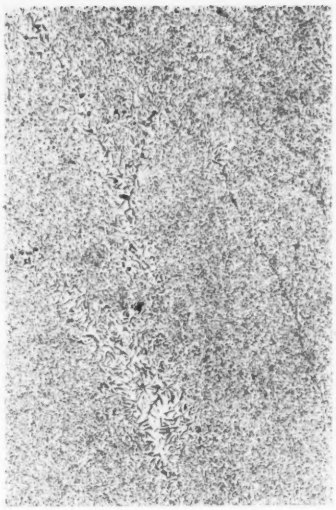

FIG. 53—Central zone in ingot 16. The hypereutectic graphite has floated to the top of the ingot. Unetched × 60

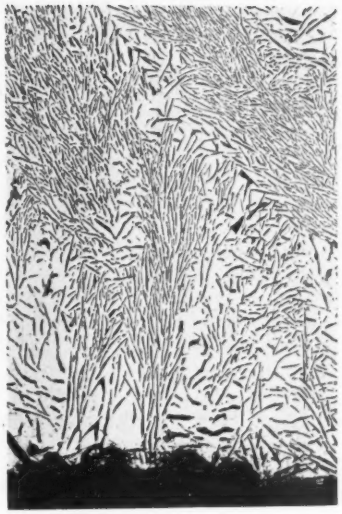

FIG. 54—Coarser graphite near the silica thermocouple sheath in ingot 16. Unetched × 60

FIG. 55—Patch of undercooled graphite in ingot 18. Unetched × 60

FIG. 56—General structure of ingot 18. Nodular graphite in metallic matrix. Etched with neutral ferric chloride × 60

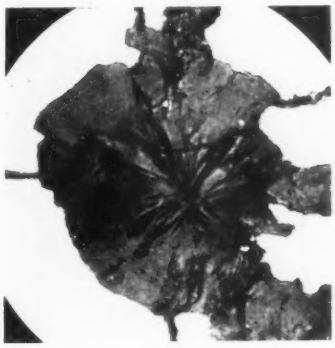

FIG. 57—Spherulitic graphite nodule in ingot 18. Under plane-polarized light. Unetched × 1500

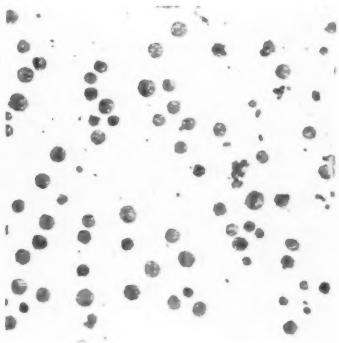

FIG. 58—General structure of ingot 20, treated with calcium silicide. Well-formed graphite spherulites. Unetched × 60

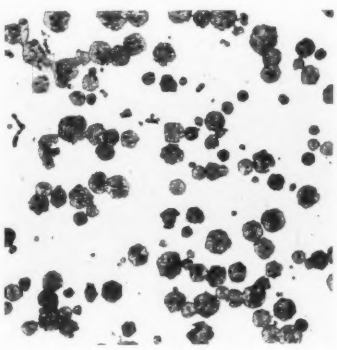

FIG. 59—Segregation of hypereutectic spherulites at the top of ingot 20, treated with calcium silicide. Unetched × 60

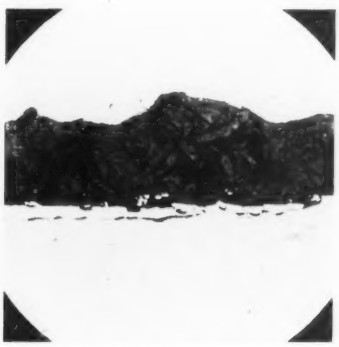

FIG. 60—Internal structure of graphite flake in ingot 21. In position of minimum reflex pleochroic absorption. Under plane-polarized light. Unetched × 1500

FIG. 61—Same spot as in Fig. 60, after rotating the specimen through 90°. Graphite flake in position of maximum reflex pleochroic absorption. Unetched × 1500

(Micrographs reduced to three-quarters linear in reproduction)

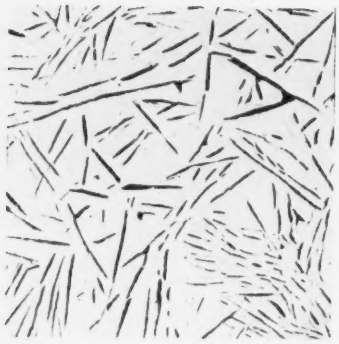

FIG. 62.—Coarse graphite structure in sulphur-treated ingot (ingot 21). Unetched × 60

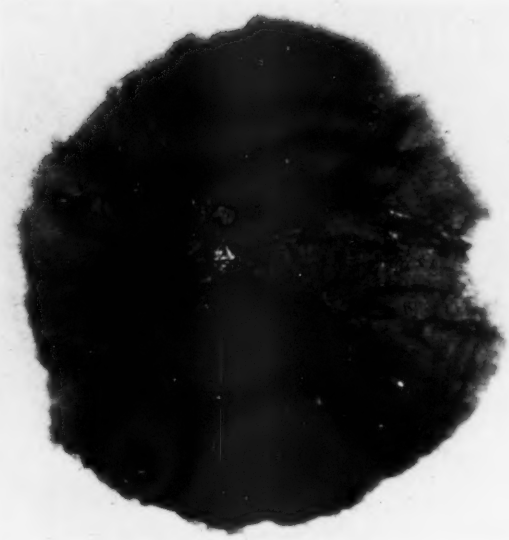

FIG. 63.—Well-formed spherulite in ingot 20. Under plane-polarized light. Unetched × 1500

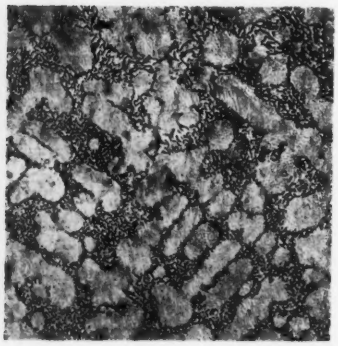

FIG. 64.—General structure of material refined with ferrosilicon-zirconium addition and containing inverse chill. (*See* Fig. 65.) Undercooled graphite in pearlite. Etched with picric acid × 150

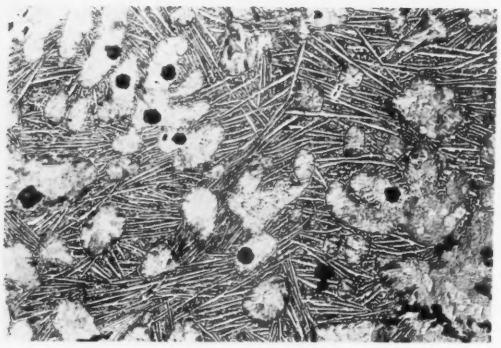

FIG. 65.—An area of inverse chill in same specimen as Fig. 64. Nodular graphite in transformed austenite dendrites with laminated eutectic. Etched with picric acid × 150

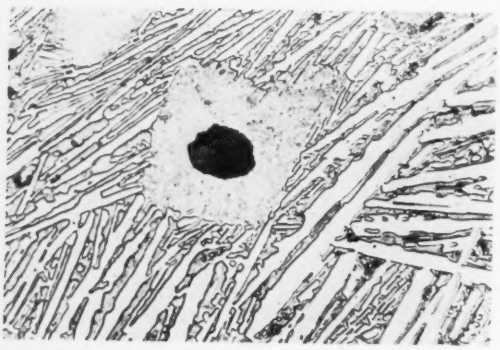

FIG. 66.—Detail of the white-iron structure shown in Fig. 65. Etched with picric acid × 600

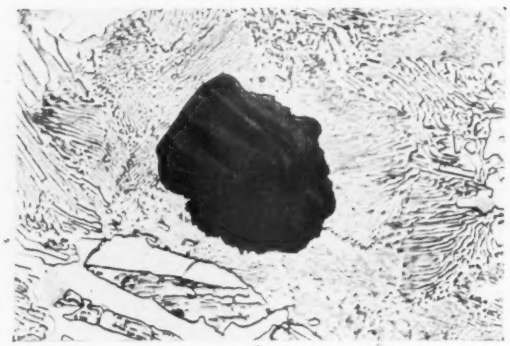

FIG. 67.—Graphite nodule in region of inverse chill shown in Figs. 65 and 66. Etched with picric acid × 1500

(Micrographs reduced to three-quarters linear in reproduction)

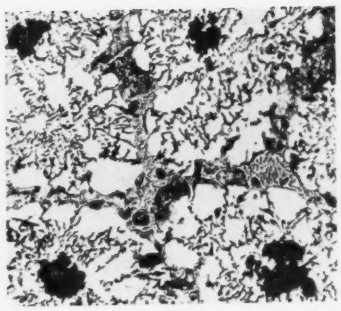

FIG. 68—Graphite nodules at centre of undercooled eutectic cells. × 200

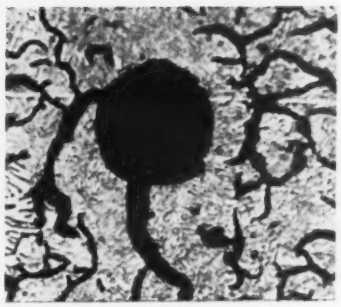

FIG. 69—Graphite spherulite with attached flake formations in hypereutectic iron. × 2000

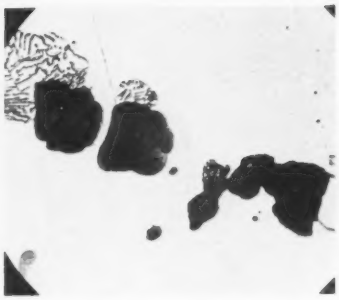

FIG. 70—Graphite spherulites in iron treated with CaSi and sulphur. Unetched × 2000

FIG. 71.—Coarse pseudo-Widmanstätten structure in iron treated with CaSi and sulphur. Unetched × 2000

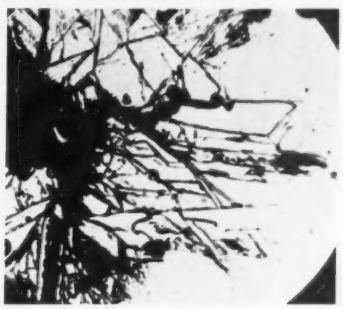

FIG. 72—Very fine pseudo-Widmanstätten structure in same sample as Fig. 71. Unetched × 2000

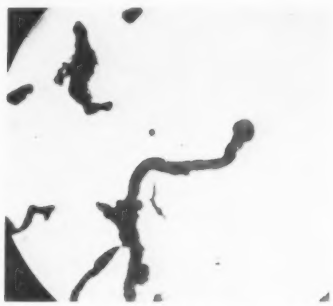

FIG. 73—Very fine spherulite formation attached to flake graphite in iron treated with CaSi and sulphur. Unetched × 2000

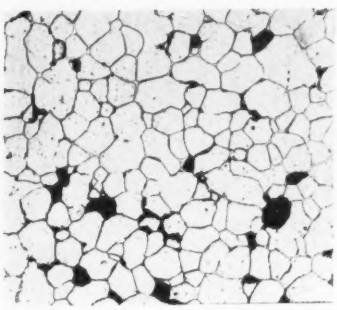

FIG. 74—Nickel–carbon alloy after attempted desulphurization with sodium carbonate. Unetched × 60

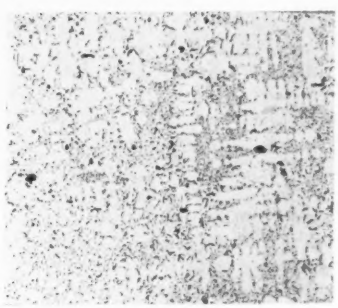

FIG. 75—Melt 31. Undercooled graphite in ingot from untreated nickel–iron–carbon alloy. Unetched × 60

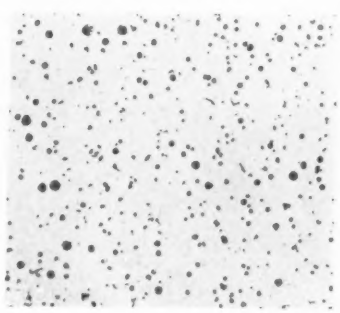

FIG. 76—Melt 31. Nodular graphite in ingot from CaSi treated nickel–iron–carbon alloy. Unetched × 60

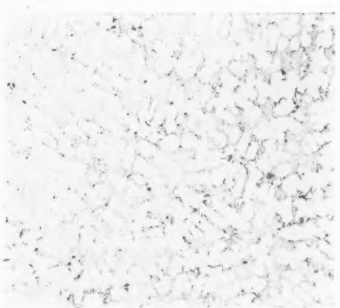

FIG. 77—Melt 32.—Undercooled graphite in ingot from untreated nickel–iron–carbon alloy. Unetched × 60

FIG. 78—Melt 32. Flake graphite in ingot from CaSi treated nickel–iron–carbon alloy. Unetched × 60

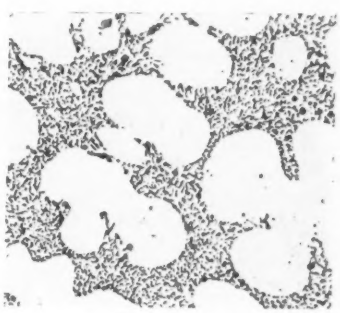

FIG. 79—Melt 33. Undercooled graphite in slowly cooled cast-iron ingot. Unetched × 150

Figs. 68 and 69 etched with picric acid

(Micrographs reduced to three-quarters linear in reproduction)

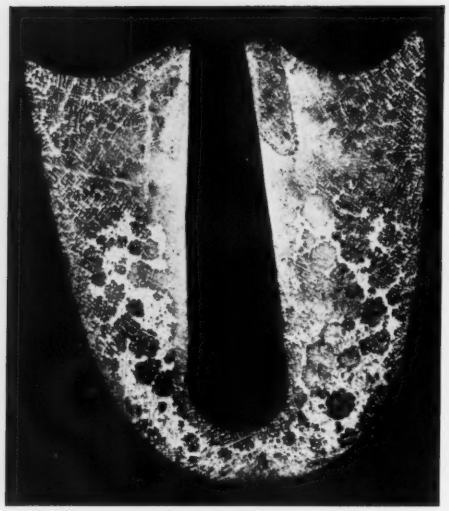

FIG. 80—Vertical section through quenched ingot (ingot 34). Light areas are white iron. Etched with picric acid × 3½

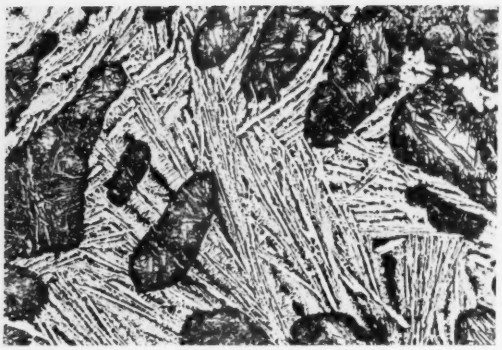

FIG. 81—Area of white iron in quenched ingot (ingot 34). Etched with picric acid × 150

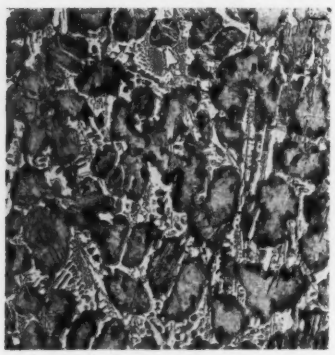

FIG. 82—White-iron structure of ingot of same material as in Fig. 81, but with chromium addition. Etched with picric acid × 60

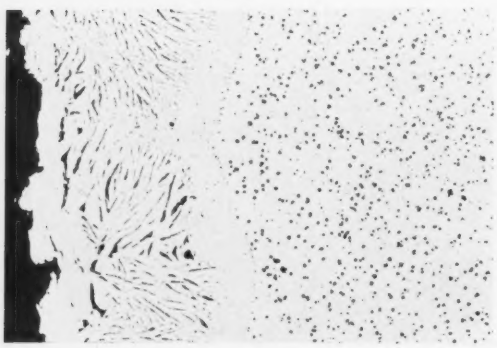

FIG. 83—Structure of nickel-carbon ingot melted in graphite crucible. Quenched just after solidification. Unetched × 60

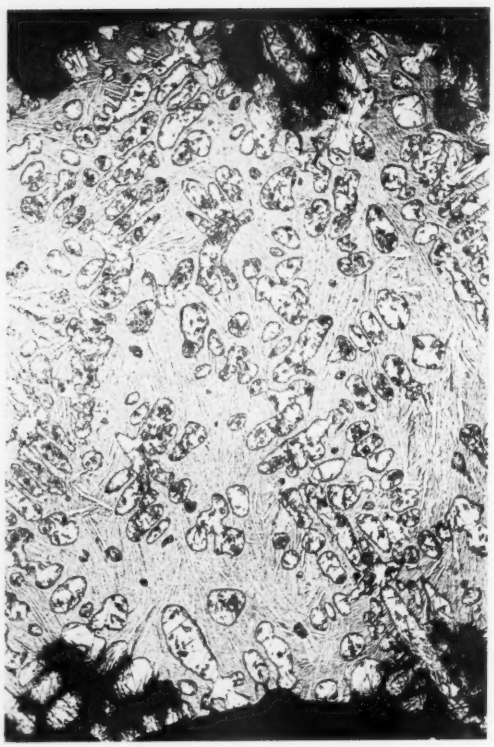

FIG. 84—White-iron structure of ingot quenched immediately after the cessation of the eutectic arrest. Etched with picric acid × 60

(Micrographs reduced to three-quarters linear in reproduction)

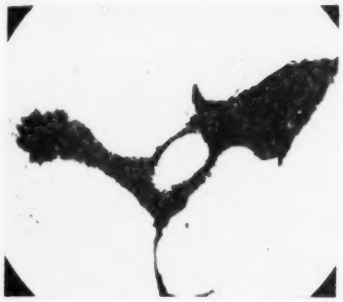

FIG. 85—Interdendritic-like graphite in ingot water-quenched from liquid condition. Unetched × 1500

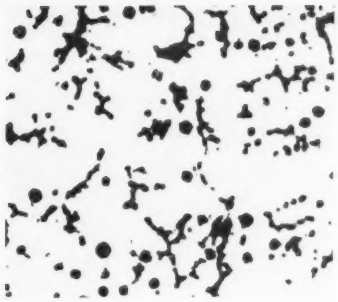

FIG. 86—Treated ingot quenched halfway through the eutectic arrest. Unetched × 60

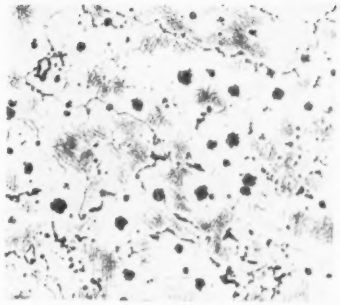

FIG. 87—Graphite nodules and fine-Widmanstätten structure of nickel carbide in water-quenched ingot. × 60

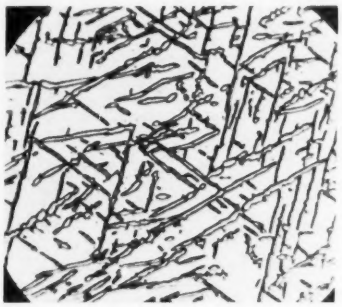

FIG. 88—Nickel-carbide Widmanstätten structure. Same sample as in Fig. 87. × 1500

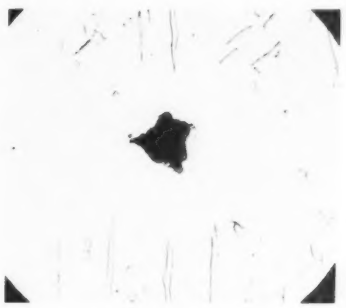

FIG. 89—Same sample as in Figs. 87 and 88, showing absence of nickel carbide in regions surrounding graphite particles. × 500

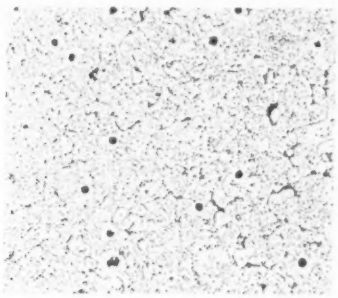

FIG. 90—Same sample as in Figs. 87 to 89, after annealing. × 60

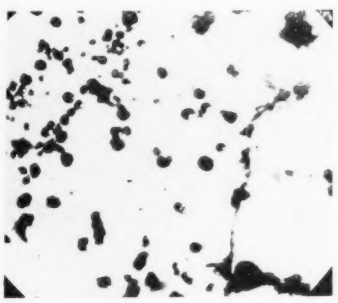

FIG. 91—Same as Fig. 90, showing fine graphite particles resulting from decomposition of nickel carbide during annealing. × 1500

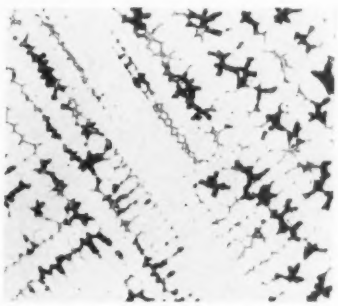

FIG. 92—Structure of quenched ingot containing approximately 60% of Fe. Carbide and interdendritic graphite. × 60

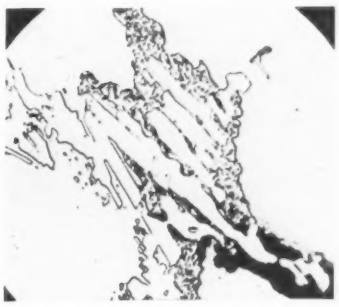

FIG. 93—Same sample as in Fig. 92, showing a eutectic carbide area. × 1500

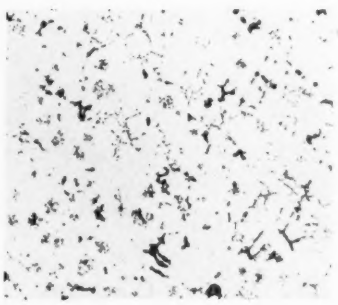

FIG. 94—Structure of quenched ingot containing approximately 40% of Fe. No carbides visible. × 60

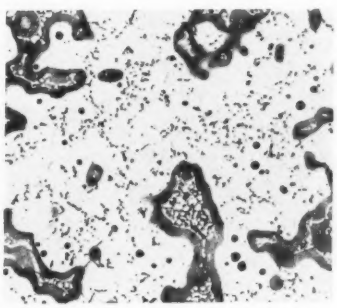

FIG. 95—Structure of CaSi treated nickel-silicon-carbon ingot. × 60

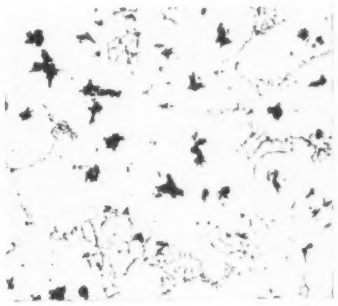

FIG. 96—Structure of Ba–Pb treated nickel-carbon ingot. Unetched × 60

Figs. 87, 88, and 90 to 95 etched with neutral ferric chloride

(Micrographs reduced to three-quarters linear in reproduction)

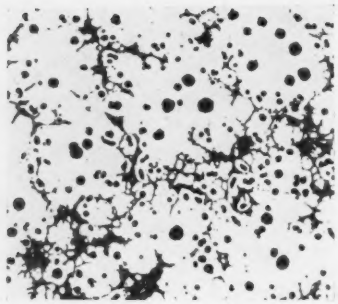

FIG. 97—Structure of Ca–Mg treated nickel–carbon ingot. × 60

FIG. 98—Same sample as in Fig. 97; $MgNi_2$ is the dispersed phase. × 1500

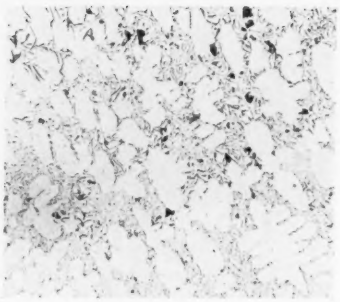

FIG. 99—Graphite structure in air-cooled cobalt–carbon melt. Unetched × 60

FIG. 100—Cobalt sulphide in same sample as Fig. 99. Unetched × 1500

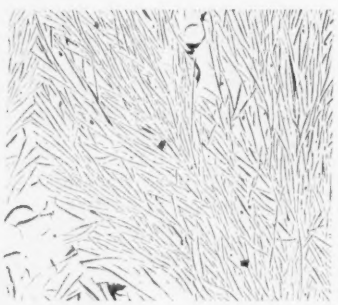

FIG. 101—General graphite structure in untreated cobalt–carbon ingot. Unetched × 60

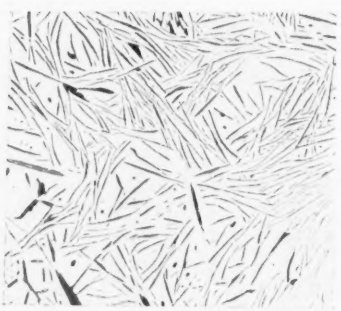

FIG. 102—General graphite structure in treated cobalt–carbon ingot. Unetched × 60

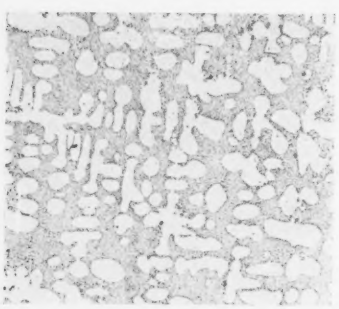

FIG. 103—Graphite area in air-cooled cobalt–carbon alloy prepared from pure cobalt powder. Unetched × 60

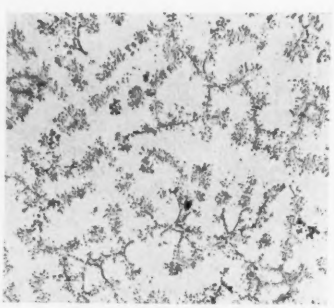

FIG. 104—Same sample as in Fig. 103, showing interdendritically arranged very fine spherulites. Unetched × 60

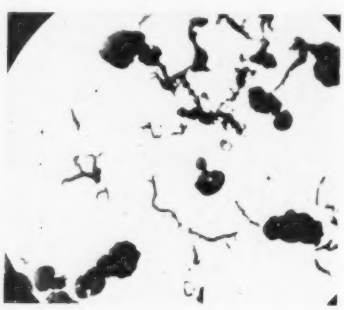

FIG. 105—Very fine spherulites in area shown in Fig. 104. Unetched × 1500

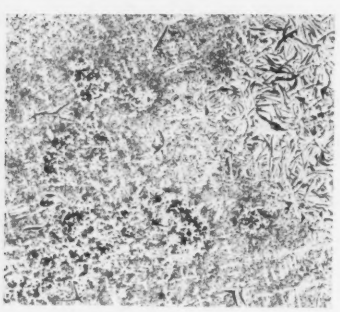

FIG. 106—Very fine undercooled graphite, some flake graphite and a few nodules in quenched cobalt–carbon ingot. × 60

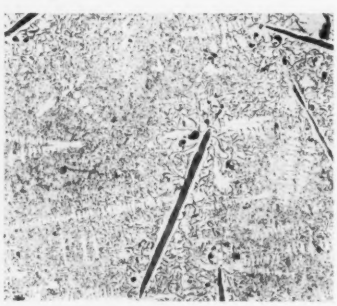

FIG. 107—Structure of quenched cobalt–carbon ingot. × 60

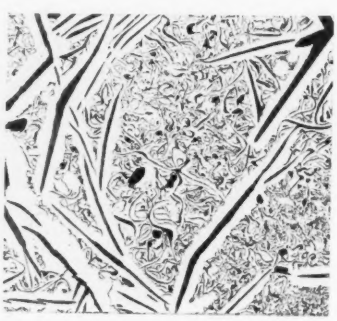

FIG. 108—Structure of quenched cobalt–carbon ingot. × 60

Figs. 97, 98, and 106 to 108 etched with neutral ferric chloride
(Micrographs reduced to three-quarters linear in reproduction)

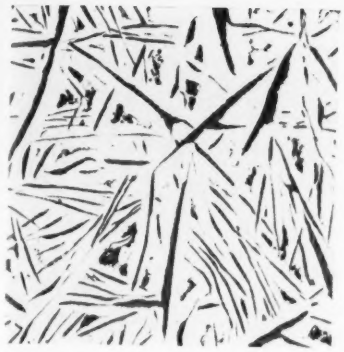

FIG. 109—Structure of quenched cobalt–carbon ingot. Etched with neutral ferric chloride × 60

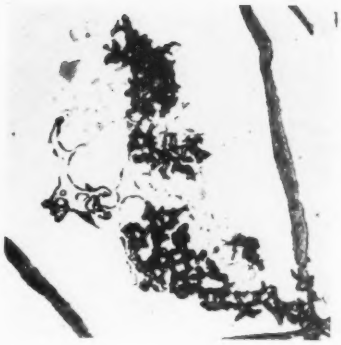

FIG. 110—Same sample as in Fig. 109, showing cobalt carbide associated with fine graphite. Etched with neutral ferric chloride × 600

FIG. 111—Massive eutectic cobalt carbide and Widmanstätten carbide in specially treated ingot. Etched with neutral ferric chloride × 1500

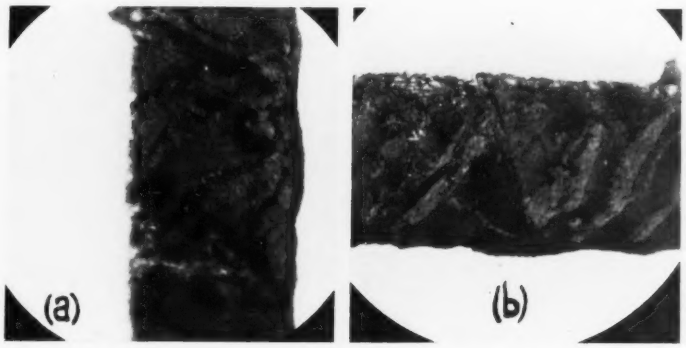

FIG. 112—Internal structure of graphite flake in a cobalt–carbon alloy: (*a*) With most of flake in position of minimum pleochroic absorption; (*b*) with most of flake in position of maximum pleochroic absorption. Under plane-polarized light. Unetched × 1500

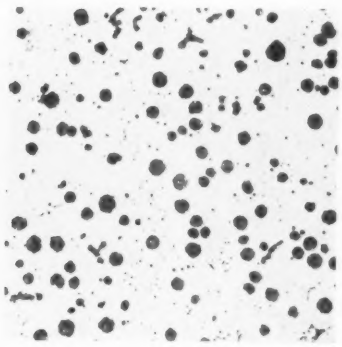

FIG. 113—Completely nodular structure of Ca–Mg treated cobalt–carbon ingot. Unetched × 60

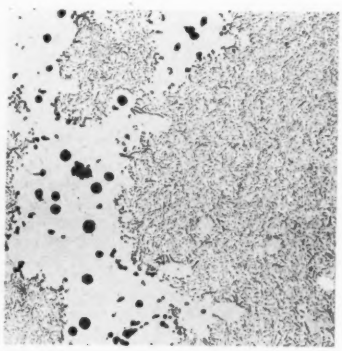

FIG. 114—Nodules and undercooled graphite in Ca–Cu treated cobalt–carbon ingot. Unetched × 60

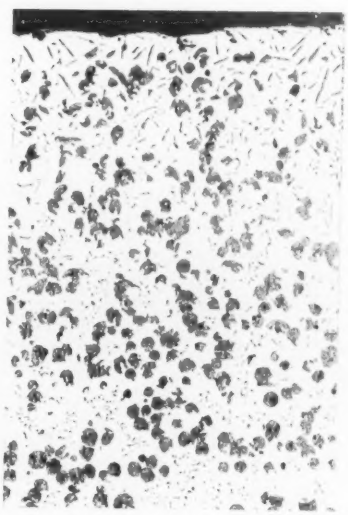

FIG. 115—Segregation of nodules at top of Ca–Cu treated cobalt–carbon ingot. Same ingot as in Fig. 114. Unetched × 60

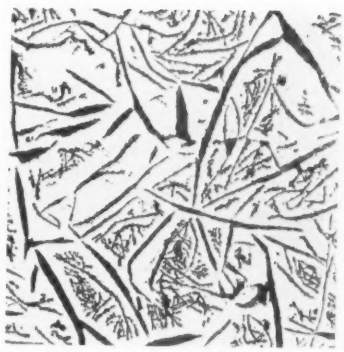

FIG. 116—Coarse graphite in Ca–Pb treated cobalt–carbon ingot. Unetched × 60

(Micrographs reduced to three-quarters linear in reproduction)

XUM

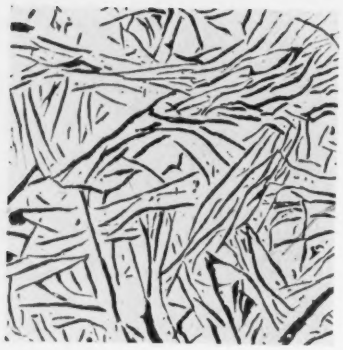

FIG. 117—Coarse flake graphite in Sr–Mg treated cobalt–carbon ingot. Unetched × 60

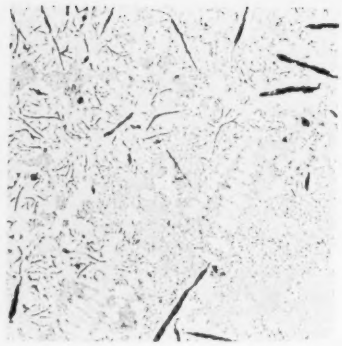

FIG. 118—Hypereutectic and fine undercooled graphite in MoS_2 treated cobalt–carbon ingot. Unetched × 60

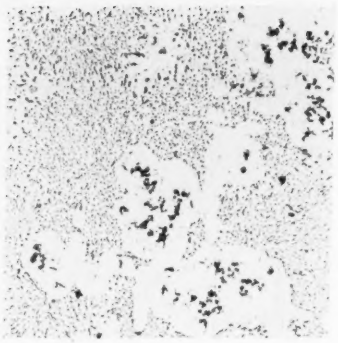

FIG. 119—"Mesh" graphite in tellurium-treated cobalt–carbon ingot. Unetched × 60

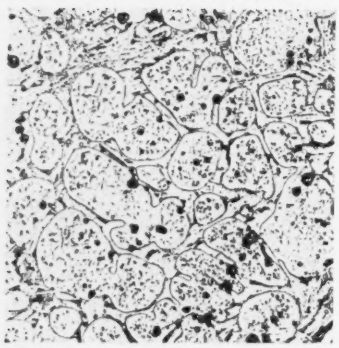

FIG. 120—Carbide and δ ferrite in quenched high-silicon cast iron. Etched with HF/HNO_3/glycerol reagent × 60

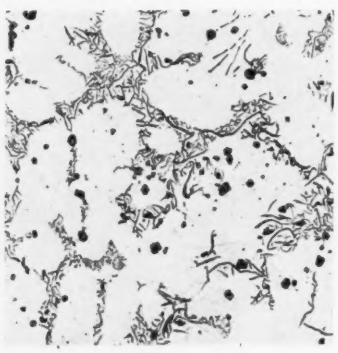

FIG. 121—Undercooled graphite + δ ferrite in quenched high-silicon cast iron. Etched with HF/HNO_3/glycerol reagent × 60

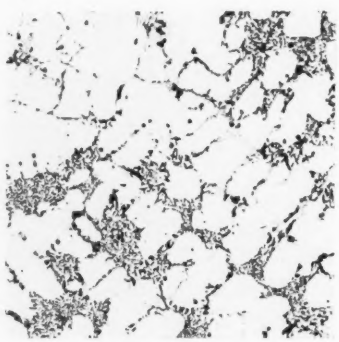

FIG. 122—Undercooled graphite and carbides in austenite matrix in good Nicrosilal casting. Etched with 10% alcoholic HCl × 60

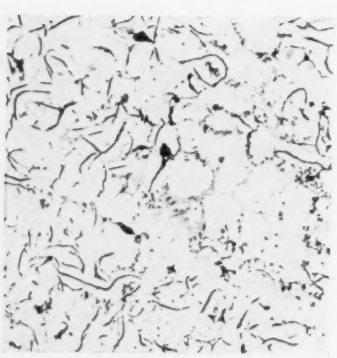

FIG. 123—Flake graphite and carbides in austenite matrix in inferior Nicrosilal casting. Etched with 10% alcoholic HCl × 60

FIG. 124—Same specimen as in Fig. 122, showing acicular dispersed carbides. Etched with 10% alcoholic HCl × 600

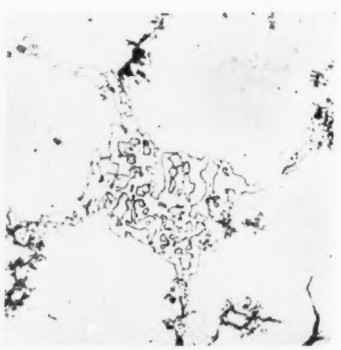

FIG. 125—Same specimen as in Fig. 123, showing relatively massive carbide. Etched with 10% alcoholic HCl × 600

(Micrographs reduced to three-quarters linear in reproduction)

calcium silicide are largely ineffectual in nucleating the melt to give flake graphite. With flake-graphite or inoculated irons solidifying at rates within the range in which flake graphite is obtained, the number of eutectic cells varies with the cooling rate. Thus within this range, and assuming that some slight undercooling takes place (most cooling curves show some slight indications of undercooling even when flake graphite is forming), then below the normal liquidus there appears to exist a range of temperature such that the number of centres or nuclei at which crystallization can begin increases with decrease in temperature. Alternatively, the nuclei become more effective at lower temperatures within this range. For cooling rates which extend the degree of undercooling beyond this temperature, a new condition applies where the number of nuclei either decrease or they become ineffectual, and in spite of an inoculating treatment the iron will solidify white or with an undercooled graphite structure, depending upon the summation of the intrinsic graphitizing and carbide stabilizing effects of the normally alloyed elements.

Influence of Hydrogen

The emphasis which Boyles placed on the effect of hydrogen in producing coarse graphite is noteworthy. Treatment with hydrogen in his experiments also increased the chilling tendency in a well-marked way, and this is a general experience. It has been argued that the effect of superheating in producing an increased chill is due to an increase in dissolved hydrogen at elevated temperatures. This involves an apparent paradox, as superheating produces undercooled graphite structures whereas hydrogen produces flake graphite structures. The hydrogen taken up by superheating would presumably originate from atmospheric moisture. Cowan [53] has demonstrated that in cupola melting the hydrogen content of the metal varies with the moisture content of the blast and that the depth of chill varies with atmospheric moisture.

Formation of Undercooled Graphite—The Work of Eash

Eash[43] has been able to show that undercooled graphite structures are formed by the decomposition of white-iron structures immediately after solidification. A bar 6 in. long by 1·2 in. in dia. was cast in a dry-sand core with a thermocouple inserted in the cavity. Cooling curves were taken on each bar cast, and, when the desired temperature had been reached, the core was broken and the specimen quenched in water. The fine graphite iron used had approximately 2·3% of total carbon and 2·6% of silicon, whilst the inoculated iron had approximately 2·1% of total carbon and 3·6% of silicon. The results obtained conclusively indicated that flake graphite irons solidify with a graphite–austenite eutectic and undercooled graphite irons solidify white, that is, with a cementite–austenite eutectic which subsequently decomposes at a temperature slightly below that indicating complete solidification. The theoretical implications of Eash's findings are immense and will be discussed in more detail later in connection with work carried out by the present authors. It is sufficient at this stage to say that, in view of the relatively crude methods used by this investigator and the somewhat abnormal compositions of his irons, the work has been repeated with the greatest care on very small cooling-curve samples for a very wide range of compositions (and for other alloys), and the findings of Eash have been fully confirmed.

Physical Properties of the Melt

Piwowarsky[41] has suggested that the variations in type of graphite structure produced by superheating may be due to variations in the molecular structure of the melt. This is a fascinating suggestion, and there are many arguments for and against the idea. Our knowledge of molecular structures in liquid metals is very limited, and the condition of carbon in cast-iron melts is unknown. Nevertheless, Schmid-Burgh, Piwowarsky, and Nipper[54] devised an ingenious apparatus to test the theory. The electrical resistivity of a ring-shaped bath of metal was measured for white and grey irons cooling from 1600° C., using a method which did not involve making any electrical connections to the melt. With grey irons a sharp increase in specific resistance was obtained during eutectic solidification which did not occur in white irons. No micrographs were given to indicate whether any of the irons had undercooled graphite, but from these results one might expect a sharp increase in resistivity immediately after solidification. The increase in resistivity of the grey irons exactly corresponded with the arrest of the accompanying cooling curve which took place at about 1130° C., which is approximately the eutectic temperature for a flake graphite iron. The investigators concluded that the results were fruitless as far as the original hypothesis was concerned. Nevertheless, the present authors feel that this method of investigation should be applied at the earliest possible moment to cast irons giving undercooled structures, and to other alloys such as aluminium–silicon, nickel–carbon, etc.

XUM

Effect of Carbon Monoxide—The Work of Schneble and Chipman

Schneble and Chipman[55] have investigated the factors involved in the superheating of grey cast iron and have attempted to explain superheating effects in terms of the gas content of the melt, particularly carbon monoxide. They constructed an induction-melting unit so that the charge could be melted and cast *in vacuo* or in any controlled atmosphere. Melts were superheated to and held at various temperatures for varying periods. In general, they found that nitrogen had little effect, that hydrogen gave flake graphite, and that carbon monoxide promoted the occurrence of undercooled graphite. This is a valuable contribution, but in attempting to explain the superheating phenomena in terms of the carbon monoxide content of the melt certain assumptions were made at variance with the known facts. They based their approach to the problem on the findings of Honda and Murakami[56] and Hayes and Scott,[57] who showed that carbon monoxide and carbon dioxide increase the graphitizing rate of solid cementite, the suggested mechanism being indicated by the following two reactions :

$$2CO = CO_2 + C \text{ (graphite)}$$
$$Fe_3C + CO_2 = 2CO + 3Fe.$$

Further, they assume that the size of the graphite flakes depends upon the stability of the carbide, so that fine graphite is produced from less stable and coarse graphite from more stable carbide. These investigators ignore the findings of Eash that the former condition results from the decomposition of solidified white iron and the other directly from the melts. The stability of cementite is known to be reduced by alloyed silicon, nickel, copper, etc., but these elements do not influence the facility with which undercooled structures can be produced by superheating. The fact that superheating also increases the tendency to chill was not referred to. That carbon monoxide can be an important factor in superheating is undeniable, but the theory of these investigators seems untenable. They give no indication of why titanium and zirconium promote undercooling without any particular superheating.

The very frequent occurrence of ferrite with undercooled graphite seems to present itself as a problem according to some references, and Piwowarsky[58] has erected some very improbable theorizing to explain its presence. He assumed this ferrite to be formed immediately after solidification. The present investigators have quenched large numbers of samples with fine undercooled graphite immediately after solidification and from lower temperatures, and have never found any ferrite to occur until the critical range has been reached. It is not generally appreciated that the state of division of graphite arising at the eutectic has a very pronounced effect in determining whether the eutectoid transformation will take place according to the stable or the metastable system, and upon the subsequent stability of any pearlite which forms on exposure to sub-critical temperatures. Taking an extreme case, an iron containing 4% of silicon and 2·8% of total carbon can be made completely ferritic when the graphite is entirely of the fine undercooled variety, and completely pearlitic when the graphite is in the normal flake form.

Carbide Stabilization

Considerable obscurity centres around the reasons why some elements are carbide stabilizers and others graphitizers. These problems are relevant to the present discussion, and a statement of the more obvious points is made below. No satisfactory explanation is known for the observed facts.

Silicon—A graphitizing element. Up to about 6% silicon progressively reduces the stability of cementite. In the range 12–14% of silicon the carbide which occurs seems to possess a stability higher than one would expect if the effect was merely an extrapolation of that at lower silicon contents. A very stable carbide (SiC) is known, soluble in cast-iron melts.

Nickel—A mildly graphitizing element. Will dissolve in cementite to give $(Fe,Ni)_3C$.[59] Ni_3C is probably very unstable. (This will be dealt with later.)

Copper—Mildly graphitizing. Similar in effect to nickel. No information about solubility in Fe_3C.

Aluminium—Graphitizing when added as a ladle addition in small amounts (less than 0·1%). This is probably owing to a deoxidizing effect. Graphitizing when added as an alloying element in amounts between 0·5 and 3%. Carbide-stabilizing in amounts ranging from 5 to 9%, probably owing to the formation of an iron–aluminium carbide.

Manganese—A carbide stabilizer. $(Fe,Mn)_3C$ probably formed. $Mn_{23}C_6$ and Mn_7C_3 have been reported in alloys with more than 80% of manganese.[60]

Chromium—A carbide stabilizer. $(Fe,Cr)_3C$ probably formed. Gives defined stable carbides. Cr_4C and Cr_3C_2, and $(Cr,Fe)_4C$, $(Cr,Fe)_7C_3$, and $(Cr,Fe)_3C_2$ probably formed.[61]

Molybdenum — A carbide stabilizer. $(Fe,Mo)_3C$ probably formed. Gives a stable carbide, probably Mo_2C, and $(Fe,Mo)_{23}C_6$ has been reported in steels.[62]

Vanadium—A carbide stabilizer. $(Fe,V)_3C$ probably formed. Gives stable vanadium carbides.

Tungsten—Not clearly defined. May be mild carbide stabilizer. Forms stable tungsten carbides WC and W_2C.

Boron—A carbide stabilizer. Forms a stable boron carbide.

Sulphur—As iron sulphide, sulphur is a carbide stabilizer. Combination with cementite is not proven. As manganese sulphide it behaves neutrally.

Tellurium—Exists as FeTe and $MnTe_2$. Appears to be a powerful carbide stabilizer in either form.

Titanium—A graphitizer when added in small amounts (less than 0·1%); this may be a deoxidizing effect resembling that of aluminium. Forms a very stable carbide (TiC), but behaves neutrally to cementite as an alloying element.

Zirconium—Behaves very similarly to titanium. Forms a stable carbide (ZrC).

Bismuth—A carbide stabilizer, in which respect it behaves similarly to tellurium, giving the same sort of rimmed chill.

It is understood that iron carbide may not be thermodynamically stable, but conventional nomenclature uses "carbide stabilization" to indicate an influence increasing the persistence of cementite, and "graphitizer" to indicate an influence decreasing its persistence. Those elements (Mn, Cr, Mo, V, and probably W) which form stable carbides and whose atoms may behave vicariously with Fe in Fe_3C, are carbide stabilizers. The action of the graphitizers (silicon, copper, and nickel) is unknown. The carbide stabilizers sulphur, selenium, and tellurium present a separate group, and we might expect a common explanation for all three. The effect of selenium has not yet been very thoroughly explored, but there are significant differences between the effects of sulphur and tellurium. Sulphur behaves as a carbide stabilizer only when in the form of FeS, but tellurium exerts this effect when present as either FeTe or $MnTe_2$ (the reasons for assigning these formulæ will be dealt with in a separate paper). Sulphur in the form of MnS behaves neutrally to cementite. The powerful effect of tellurium no doubt prompted its commercial application.[63] In addition to this difference, tellurium imparts a characteristic appearance to the chill, *viz.*, a very sharp line of demarcation between white and grey zones. Morgan, working in the British Cast Iron Research Association laboratories, found the chilling effect of tellurium to be very sensitive to pouring temperature.[64] Fig. 27 shows the effect of pouring temperature on the fracture of an iron containing tellurium. It might be suggested that this apparent influence of tellurium was due to its gradual loss from the melt. This is disproved by the fractures illustrated in Fig. 28. In this case an iron treated with tellurium was remelted in a crucible furnace and then poured at varying temperatures. The same effect will again be noted. There undoubtedly has been some loss of tellurium during remelting, but even after this lengthy process sufficient has remained to give a marked effect and to cause the sharp chilled rim to return with the low pouring temperature. The effect of pouring temperature on a characteristic sulphur chill is shown in Fig. 26. Figs. 26, 27, and 28 were all taken from hitherto unpublished work by Morgan, and all show the fractures of 1·2-in. test-bars poured in green sand. It is quite obvious from these illustrations that, whatever factors cause chill depth to increase with increasing pouring temperature, tellurium increases the sensitivity of the iron to these influences. In sulphur and tellurium a very complex case of carbide stabilization exists. From work not yet complete, it appears that bismuth may behave in a manner similar to that of tellurium.

The graphitizing influence of aluminium, titanium, and zirconium, when added in small amounts immediately before pouring, may be related to their "deoxidizing" effects. There is no sure foundation for this assumption, but on the face of it it seems to be the only property which they all possess in common. There is ample ground here for investigation of the possibility of oxygen content and the method of its fixation influencing chilling characteristics. The advantages occurring from the use of aluminium along with a ferrosilicon inoculant and of the use of a ferrosilicon containing manganese and zirconium (S.M.Z.) have been referred to previously. A significant fact in this context is that whilst these two give, on the average, higher-strength grey irons than those obtained by using ferrosilicon alone, no differences in the graphite or matrix structure of grey material can be observed (irons treated with S.M.Z. always contain the characteristic zirconium sulphides). The authors have no leading observations to make upon the carbide-stabilizing effect of boron. Boron added in small amounts has been reported[65] to have a graphitizing effect when added to

malleable iron. This effect has also been noted in the laboratories of the British Cast Iron Research Association when ferroboron was used for the alloy addition. Ferroboron usually contains large amounts of aluminium or other deoxidizers, and it is suggested that these are responsible for the graphitizing effects. No such graphitizing effect has been noted when boron carbide was used to make the addition to grey and mottled irons. The effect of alloyed aluminium is anomalous, but it should be noted that an aluminium carbide (Al_4C_3) can be formed.

The industry will recognize other influences on carbide stabilization than those listed above. As a general rule, all other conditions being the same, the tendency to chill increases with decreasing total carbon content. Other effects, all very familiar to practical foundrymen, may be related to the oxygen content of the melt. This will include the effect of slag composition and melting stock. Compositions being equal, the chilling tendency and graphite structures of irons prepared from hematite and steel on the one hand, and refined iron on the other, will differ. Similarly, electric-furnace, crucible-melted, and cupola-melted irons may vary in structure and chilling tendency. On the whole, cupola irons will show a coarser graphite structure with less tendency to give undercooled structures than is the case with electric-furnace and crucible-melted irons. It can be argued that this is owing to an inoculating effect obtained by the metal being in contact with coke. On the other hand, the atmospheres in these melting furnaces are quite different. The practical foundryman also reports that the chilling tendency of an iron prepared by mixing molten high-silicon and molten low-silicon irons differs from that of a single melt of the same final silicon analysis. Many instances of this type obtrude upon the problem, but few have been stated with sufficient precision to enable them to be considered in detail in the present discussion.

Effect of Slag Composition

Particularly revealing were the experiments of Bardenheuer and Reinhardt,[95] who found that melting under mill scale in a high-frequency induction furnace tended to produce flake graphite, and melting under glass tended to produce undercooled graphite. The whole field of slag–metal reactions for cast iron is practically unexplored, but this example suffices to show the importance of its consideration. The acidity or basicity of the slag may determine the FeO or SiO_2 content of the melt, which may in turn affect the tendency to undercool. On the other hand, it could be argued that high-silica slags react with the carbon of the melt to give carbon monoxide and silicon. The silicon would be taken up by the melt and, according to Schneble and Chipman, the carbon monoxide would facilitate undercooling. These suggestions, like many others in this field, are purely speculative.

Inverse Chill

Bound up with the mechanism of formation of undercooled graphite structures is the fact that inverse chill occurs most frequently in undercooled-graphite-containing irons. (Inverse chill refers to the fracture or microstructure which arises when the central and more slowly cooled zones of a section have a white-iron structure and the outer and more rapidly cooled zones have a grey-iron structure. This is the reverse of what one would normally expect.) No satisfactory explanation has been given for inverse chill, but this connection must throw some light on its formation. The present authors distinguish two varieties of inverse chill. In irons which have completely fine undercooled graphite structure, inverse chill occurs in a wide variety of degree, but in every case with a characteristic appearance. In small amounts it has the appearance of an undissolved alloy addition, and under the microscope appears as indicated in Fig. 29. Note should be made of the acicular character of the carbides in this micrograph. Fig. 30 indicates typical fractures of bars with undercooled graphite and central areas of inverse chill. In a later section of this paper this matter is dealt with again. Another type of inverse chill occurs when the grey portions of the melt are not necessarily of the undercooled structure. In these cases the outer grey portions are generally only a very thin band (Fig. 31). It is suggested that this is an inoculating effect of the mould face. In a similar manner, when deliberately producing irons with undercooled graphite structures, a fine rim of normal flake graphite has been found at the edges of the test-bars, again apparently owing to the silica sand or coal dust at the mould face inoculating the casting during solidification. In another context, silica will be shown to have an inoculating effect, and, using the theory of Schneble and Chipman, it is conceivable that under these conditions silicon reacts with the carbon of the melt to give carbon monoxide, which in turn facilitates graphite formation. In this latter case the effect would not be a true inoculating effect.

Reference has been made to the carbide-stabilizing effect of bismuth, but this element exercises another influence of interest; it has

a pronounced graphite-refining effect. Treatment of a melt with bismuth is as effective as titanium in replacing flake graphite structures with undercooled graphite structures, even without any subsequent treatment of the melt with carbon dioxide. Figs. 32 and 33 illustrate this effect in 1·2-in. test-bars, the untreated metal having the following analysis :

T.C.	Si	Mn	S	P
3·18%	1·89%	0·63%	0·034%	0·045%

An addition of 0·2% of bismuth was made to the iron shown in Fig. 33, but chemical analysis gave only 0·02%.

Summary of Graphite-Formation Anomalies

A detailed account has now been given of the more important anomalies of graphite formation in grey cast iron, and it is appropriate at this stage to recapitulate with a short summary. The outstanding factors needing explanation, correlation, or proof may be listed as follows :

(1) Formation of undercooled structures.
(2) Production of undercooled structures by superheating.
(3) Production of undercooled structures by titanium and zirconium, with or without carbon dioxide.
(4) Effect of carbon dioxide in (3).
(5) Production of undercooled structures by bismuth.
(6) Titanium and zirconium refine only when present as TiC and ZrC.
(7) Von Keil's silicate-slime theory.
(8) Norbury and Morgan's theory.
(9) Removal of porosity by undercooled graphite structures.
(10) Inoculating effect of ferrosilicon and calcium silicide.
(11) Inoculating effect of graphite.
(12) Improvement in ferrosilicon inoculation by inclusion of powerful deoxidizers such as zirconium or aluminium.
(13) Explanation of undercooling according to classical theory.
(14) Simultaneous occurrence of primary dendrites from both sides of eutectic.
(15) Inoculating effect of kish.
(16) Theory of superheating effect being due to elimination of graphite nuclei (Massari and Lindsay).
(17) Rapid solution of graphite above the liquidus (Piwowarsky).
(18) Inoculation effect of ferrosilicon owing to the concentrations of silicon giving graphite (Eash).
(19) Formation of undercooled structures by decomposition of a white iron (Eash).
(20) Effect of hydrogen : Flake graphite and carbide stabilizer.
(21) Vacuum melting : Fine undercooled graphite.
(22) Effect of sulphur : Flake graphite.
(23) Large eutectic cell size of undercooled structures.
(24) Flake graphite forming directly from melt.
(25) Effect of moisture in cupola blast in increasing chill.
(26) Effect of carbon monoxide (Schneble and Chipman).
(27) Effect of increasing pouring temperature in increasing chill.
(28) Effect of superheating in increasing chill.
(29) Carbide-stabilizing action of chromium, manganese, carbon, molybdenum, and tungsten.
(30) Carbide-stabilizing action of sulphur, selenium, and tellurium.
(31) Graphitizing action of silicon, nickel, and copper.
(32) Graphitizing action of ladle additions of aluminium, titanium, and zirconium.
(33) Carbide-stabilizing action of bismuth.
(34) Carbide-stabilizing action of boron.
(35) Aluminium first a graphitizer (up to about 3%), then a carbide stabilizer.
(36) Sharp chill with tellurium, and its sensitivity to pouring temperatures.
(37) Formation of inverse chill.
(38) Effect of melting unit.
(39) Effect of nature of melting stock.
(40) Refining action on graphite of copper in excess of liquid miscibility.
(41) Effect of slag composition.

Experiments on the Nickel–Carbon Alloys

Object of this Work

At the outset these experiments were motivated by the feeling that the findings of Eash that undercooled fine graphite structures were formed by the decomposition, immediately after solidification, of a white-iron structure were the most important in recent years. With this knowledge it was obvious that the picture of undercooling based on our knowledge of the betol–salol system was at the best limited, and certainly unsatisfactory. In the investigations on undercooling in this system there is no reference to the formation of metastable compounds formed between betol and salol at the hypertectic temperature. Therefore the question arose as to whether the mode of undercooling in iron–carbon–silicon alloys and cast irons was unique, or whether

this type of solidification could be reproduced for other systems, by the study of which we might be led closer to the truth about cast irons. This course was further favoured in view of the manner in which the form of the binary equilibrium diagrams of two elements with a third element can be related to their positions in the periodic classification of elements. There is frequently considerable analogy between two systems, *AB* and *AC*, when *A* is an element common to both, and *B* and *C* are elements of the same group in the periodic table of elements. Useful inferences about the system *AB* may be obtained from a study of the system *AC*, and *vice versa*. Inferences of this type must, of course, be accepted with extreme caution, and the authors have tried to preserve this cautious attitude throughout the following experiments.

When considering the problem of graphite formation in iron alloys, the value of studying the nickel–carbon and cobalt–carbon systems came immediately to mind. Iron, cobalt, and nickel are neighbours in the first long period and in this period they are the last three transitional elements. Their atomic numbers are 26, 27, and 28, and their atomic weights are 55·85, 58·94, and 58·69, respectively. The increase in electrons consequent upon increasing atomic number takes place by the addition of electrons to the inner orbits.

With carbon, these three elements all yield equilibrium diagrams of great similarity. The systems are eutectiferous, involving a eutectic of graphite and a solid solution of carbon in the metal. The crystal structure of these solid solutions is probably the same in all three cases The liquidus on the metal-rich side of this eutectic corresponds to the deposition of this solid solution from the melt, and on the carbon-rich side of the eutectic it indicates the solubility of graphite in the melt. In every case the solubility of carbon in the solid solution decreases with decreasing temperature.

It was decided to investigate the nickel–carbon alloys first. This decision was taken largely because a supply of reasonably pure nickel was readily available. In addition, this element is a much more important alloying element in cast iron than is cobalt. The nickel–carbon system has been investigated on several occasions,[67–70] and there is fair agreement in the results obtained. The diagram is given in Fig. 126, in which all the available data have been included. Ruff and Martin[68] claimed to have obtained the carbide Ni_3C, indicated by the break on the liquidus curve, by quenching from a high temperature. Their micrographs give no positive evidence of

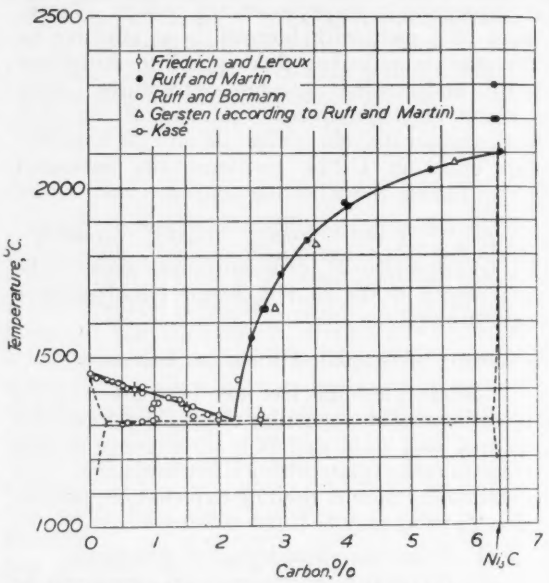

FIG. 126—Equilibrium diagram for the nickel–carbon alloys (Marsh)

this compound. The general inference to be made from this early work is that if a nickel carbide exists, it is extremely unstable and that all alloys with carbon in excess of solid solubility consist of solid solution + graphite. Jacobson and Westgren[71] prepared Ni_3C by carburizing nickel powder in carbon monoxide at approximately 285° C. for 24 to 168 hr. Their X-ray analysis of this compound showed it to be hexagonal close-packed and hence quite different from cementite.[97,98] The nickel carbide Ni_3C would appear therefore to be very unstable at high temperatures, but at least to be persistent at low temperatures.

First Series of Experiments on Nickel–Carbon Alloys

The first series of experiments on nickel–carbon alloys were all carried out in a small high-frequency induction furnace, using a fireclay crucible holding about 8 lb. of metal. The charges in every case consisted of pure nickel shot and electrode carbon. For micro-examination and chemical analysis small ingots 1·2 in. in dia. and 2–6 in. long were cast in green-sand moulds. The initial object of these experiments was to determine whether nickel–carbon alloys reacted to various treatments in a manner similar to that of cast irons. A large number of melts were carried out in this series, but, for the

sake of brevity, only the more important melts indicating some definite fact will be described. Similarly, every sample cast was examined microscopically and photographed, but here only the important relevant features will be illustrated by micrographs.

Melt No. 1—This was of hypo-eutectic composition, and the charge contained 2% of carbon. The metal bath was superheated to 1600° C. and the first ingot, 1*A*, was poured. Calcium silicide (0·5%) was then added to the melt and ingot 1*B* was poured. The melt was then treated with ferrosilicon-zirconium* (0·5%) and ingot 1*C* was poured.

Ingot 1*A* had a graphite structure exactly similar to the undercooled structure of a hypo-eutectic cast iron. The actual structure is shown in Fig. 34.

Ingot 1*B* had a structure consisting entirely of spherulitic carbon nodules. This structure is illustrated in Fig. 35. This sample resembled the structure of an annealed malleable iron, and its occurrence in a cast ingot without any heat-treatment was considered a remarkable occurrence. Ingot 1*C* had a structure consisting entirely of the fine undercooled type of graphite as found in ingot 1*A*. This structure is illustrated in Fig. 36.

Quite obviously the results of this melt indicated that superheating had a similar effect in nickel–carbon alloys to that in cast iron. Calcium silicide additions had a very different effect, and ferrosilicon-zirconium probably behaved as in cast irons.

Melt No. 2—This was carried out originally to investigate the effect of adding a cementite stabilizer to a nickel–carbon alloy. It was fully appreciated that because an element may stabilize iron carbide in iron–carbon–silicon alloys, it does not therefore necessarily have the same effect on carbides in other systems, in this case Ni_3C in nickel–carbon alloys. Examples of this kind are common in cast irons themselves; for instance, silicon has no apparent effect on the stability of titanium carbide, and chromium-containing carbides in the austenitic iron Nicrosilal are remarkably persistent in spite of the simultaneous presence of 17–20% of nickel and 4–5% of silicon, the combined effects of which should easily neutralize the carbide-stabilizing influence of 2–4% of chromium, according to the popularly accepted relative values of these influences. Many other examples can be quoted, and these instances might assist in a better understanding of the reasons for the observed effects of these elements. In this instance tellurium was chosen.

The charge was exactly the same as that used for melt No. 1. After superheating, ingot 2*A* was poured. This gave an undercooled graphite structure exactly similar to 1*A*. Tellurium (0·2%) was then added and ingot 2*B* was poured; the melt was then treated with 0·6% of ferrosilicon-zirconium and ingot 2*C* was poured. Ingots 2*B* and 2*C* both contained coarse flake graphite structures. The structure of the tellurium-treated ingot is shown in Fig. 37. This coarse graphite structure was exactly similar in appearance to its counterpart in grey cast iron. Under these conditions and in the presence of tellurium, the ferrosilicon-zirconium did not appear to have any effect. After treatment with tellurium, small amounts of a minor phase, probably nickel telluride (NiTe) [72] were observed in the structure. No carbides were seen.

Melt No. 3—This melt was carried out to determine whether the nodular structure produced by calcium silicide would persist after treatment with tellurium. The charge used was the same as those used previously. After superheating to 1600° C., an ingot, 3*A*, was poured; an addition of calcium silicide was then made and ingot 3*B* was poured. Finally, tellurium was added and ingot 3*C* was poured.

Ingot 3*A* was exactly similar to ingot 1*A*, showing the undercooled graphite type of structure (*see* Fig. 34). Ingot 3*B* had the graphite in the spherulitic form as in Fig. 35, and ingot 3*C* had coarse flake graphite as in Fig. 37.

These results indicate the predominating influence of the tellurium over the calcium silicide addition.

Melt No. 4—As tellurium has, by virtue of its position in the periodic table, similar properties to those of sulphur, and as the latter element is more important as a constituent in cast iron, it was decided to investigate its effect when present as nickel sulphide and when present as manganese sulphide.

The charge employed was similar to those used previously. After superheating to 1600° C., ingot 4*A* was poured; 0·3% of roll sulphur was then added and ingot 4*B* was poured; 1% of manganese as ferromanganese (80%) was then added and ingot 4*C* was poured.

Ingot 4*A* had the undercooled graphite structure (*see* Fig. 34). Ingot 4*B* had a coarse flake graphite structure, and ingot 4*C* was similar to 4*B*, but with slightly finer graphite. These two structures are illustrated in Figs. 38 and 39, respectively.

* It has already been shown in an earlier section that this addition causes the refinement of the graphite structure in cast iron.

The nickel sulphide, Ni_3S_2[47,73-75] was visible in ingot 4*B*, and only manganese sulphide was seen in ingot 4*C*.

In this case an obvious analogy exists between nickel–carbon alloys and iron–carbon–silicon alloys. It will be recalled from an earlier section that Boyles had demonstrated that sulphur additions to iron–silicon–carbon alloys causes the formation of flake graphite structures.

Melt No. 5—This melt was carried out to investigate the effect of hydrogen on superheated metal. A similar charge to that used previously was superheated to 1600° C. and ingot 5*A* was poured. The melt was held at this temperature, hydrogen was bubbled through for 2 min., and ingot 5*B* was poured.

The structure of ingot 5*A* is shown in Fig. 40 and consisted of a mixture of the fine undercooled variety and a coarser form. Ingot 5*B* had a uniform graphite structure rather coarser than the typical undercooled variety reported previously. This structure is illustrated in Fig. 41.

Melt No. 6—To investigate the effect of hydrogen treatment upon a melt already treated with calcium silicide, a charge similar to those previously used was superheated to 1600° C. and ingot 6*A* was poured ; calcium silicide was then added and ingot 6*B* was poured. The melt was then kept at temperature, hydrogen was bubbled through for 2 min., and ingot 6*C* was poured.

Ingot 6*A* showed the familiar undercooled graphite structure. Ingot 6*B* had all the graphite in the nodular form (as in Fig. 35), with the exception of a rim of undercooled graphite (Fig. 43). Ingot 6*C* had a structure consisting entirely of fine undercooled graphite, as illustrated in Fig. 42. Obviously, the nodular structure produced by calcium silicide can be removed by hydrogen. The undercooled graphite rim on ingot 6*B* will be referred to later.

Melt No. 7—Again the charge was similar to the previous melts. After superheating to 1600° C. ingot 7*A* was poured. Tellurium was then added and ingot 7*B* was poured ; hydrogen was then bubbled through the melt for 2 min. and ingot 7*C* was poured. Ingot 7*A* had an undercooled graphite structure and ingots 7*B* and 7*C* had coarse flake graphite structures. Hydrogen therefore had no effect on the coarsening influence of tellurium.

Melt No. 8—The charge in this case contained $3\frac{1}{4}\%$ of carbon, that is, sufficient to render the melt hypereutectic. After superheating to 1600° C., ingot 8*A* was poured ; calcium silicide was then added and ingot 8*B* was poured.

Ingot 8*A* had long straight " kish " or hypereutectic graphite flakes with an undercooled graphite eutectic, together with primary dendrites from the metal-rich side of the eutectic (Fig. 45). This may be taken to indicate undercooling of the type described previously and indicated in Fig. 7*c*. Ingot 8*B* had a structure consisting entirely of spherulitic carbon nodules (Fig. 44). It would appear that both hypereutectic and eutectic graphite can be caused to form in spherulites by the calcium silicide treatment of the melt.

Melt No. 9—The charge for this melt contained $3\frac{1}{4}\%$ of carbon, as for melt No. 8. After superheating, ingot 9*A* was poured, tellurium was then added, and ingot 9*B* was poured.

Ingot 9*A* showed areas very similar in structure to ingot 8*A*, but considerable flotation of kish graphite to the top regions of the ingot had taken place. As a result, towards the upper central zones of the melt the structure was wholly eutectic (Fig. 46), whilst at the top the graphite was much coarser and some well-formed spherulites were to be seen (Fig. 47). It would appear that hypereutectic graphite can cause inoculation of the solidifying ingot and also that this graphite can give spherulites without additions of calcium silicide. Ingot 9*B*, treated with tellurium, had a uniformly very coarse graphite structure (Fig. 48).

Melt No. 10—A charge containing 2% of carbon, that is, hypo-eutectic, was superheated to 1600° C. and ingot 10*A* was poured. Titanium (0·5%) as ferrosilicon-titanium was then added, and, when dissolved, carbon dioxide was bubbled through the melt for 2 min. ; ingot 10*B* was then poured.

Ingot 10*A* had a fine undercooled graphite structure (Fig. 49) and ingot 10*B* had a similar, but coarser graphite structure (Fig. 50).

Melt No. 11—A charge containing 2% of carbon was superheated to 1600° C. and calcium silicide was added. Ingot 11*A* was then poured. Titanium (0·5%) was then added, carbon dioxide was bubbled through the melt for 2 min., and then ingot 11*B* was poured.

Ingot 11*A* had its graphite in nodular form as illustrated previously, but in ingot 11*B* a graphite structure similar to that illustrated in Fig. 50 had formed.

Melt No. 12—A charge containing 2% of carbon was superheated to 1600° C. and then 0·5% of sulphur, as roll sulphur, was added and ingot 12*A* was poured. The melt was then treated with calcium silicide and ingot 12*B* was poured.

Ingot 12*A* and ingot 12*B* both had fairly coarse flake graphite structures. In the presence of

TABLE I—*Summary of First Series of Nickel–Carbon Melts*

Ingot No.	Charge Composition	Treatment	Structure
		Melt No. 1	
1*A*	2% C	Superheated	Undercooled
1*B*		CaSi	Nodular
1*C*		FeSiZr	Undercooled
		Melt No. 2	
2*A*	2% C	Superheated	Undercooled
2*B*		Te	Coarse
2*C*		FeSiZr	Coarse
		Melt No. 3	
3*A*	2% C	Superheated	Undercooled
3*B*		CaSi	Nodular
3*C*		Te	Coarse
		Melt No. 4	
4*A*	2% C	Superheated	Undercooled
4*B*		S	Coarse
4*C*		Mn	Coarse
		Melt No. 5	
5*A*	2% C	Superheated	Undercooled
5*B*		H_2	Coarser
		Melt No. 6	
6*A*	2% C	Superheated	Undercooled
6*B*		CaSi	Nodular
6*C*		H_2	Coarser
		Melt No. 7	
7*A*	2% C	Superheated	Undercooled
7*B*		Te	Coarse
7*C*		H_2	Coarse
		Melt No. 8	
8*A*	3¼% C	Superheated	Kish and undercooled
8*B*		CaSi	Nodular
		Melt No. 9	
9*A*	3¼% C	Superheated	Kish and undercooled
9*B*		Te	Coarse
		Melt No. 10	
10*A*	2% C	Superheated	Undercooled
10*B*		Ti + CO_2	Coarser
		Melt No. 11	
11*A*	2% C	Superheated + CaSi	Nodular
11*B*		Ti + CO_2	Coarse undercooled
		Melt No. 12	
12*A*	2% C	S	Coarse
12*B*		CaSi	Coarse
		Melt No. 13	
13*A*	2% C	S + Mn	Coarse
13*B*		CaSi	Coarse
		Melt No. 14	
14	2% C	Si + Mn + P + S	Coarse

sulphide, calcium silicide has no tendency to cause the formation of nodular structures.

Melt No. 13—A charge was melted as for melt No. 12, 0·5% of sulphur as roll sulphur was added, then 1·0% of manganese as 80% ferromanganese, and ingot 13*A* was poured. Calcium silicide was then added and ingot 13*B* was poured.

Both of these ingots had coarse flake graphite structures. Calcium silicide therefore does not give nodular structures in the presence of manganese sulphide.

Melt No. 14—The charge for this melt was adjusted to give a composition resembling that of a cast iron, but based on nickel instead of iron. It consisted of pure nickel shot, 2% of electrode carbon, 2·5% of silicon as silicon metal, 1·25% of manganese as ferromanganese, 1·3% of phosphorus as ferrophosphorus (20%), and, when molten, 0·3% sulphur was added as roll sulphur. One ingot was cast from this melt, and its structure showed very coarse flake graphite with islands of nickel phosphide (Fig. 51).

Other melts were carried out in an attempt to cause the formation of flake graphite by inoculation with ferrosilicon and electrode carbon, but no success was achieved. Attempts to remelt ingots of nickel–carbon alloys resulted in a very heavy loss of carbon, so that no eutectic graphite was visible in the resulting ingots. Similarly, attempts to desulphurize nickel–carbon melts containing sulphur by treatment with sodium carbonate caused an almost complete loss of carbon, giving ingots consisting of polyhedral grains surrounded by nickel sulphide (Fig. 74). It would appear that the carbon in nickel–carbon melts is very reactive to atmospheric oxygen and, probably, carbon dioxide.

A summary of the above melts and the inferences to be drawn from them is given in Table I.

It seems possible to draw the following tentative conclusions from this series of nickel–carbon melts :

(1) Undercooled, flake, and nodular structures, closely resembling those found in cast irons, can be reproduced at will.

(2) Additions of calcium silicide to the melt cause the formation of nodular structures.

(3) Additions of tellurium or sulphur give coarse flake graphite in all nickel–carbon alloys.

(4) In respect of sulphur additions, nickel–carbon alloys behave similarly to iron–carbon–silicon alloys.

(5) The coarsening effect of sulphur persists whether present as Ni_3S_2 or MnS.

(6) Calcium silicide additions in the presence of Ni_3S_2 or MnS have little or no effect.

(7) Treatment with hydrogen may coarsen the graphite.

(8) Treatment with hydrogen after treatment with calcium silicide prevents the formation of nodules.

(9) Treatment with titanium and carbon dioxide tends to coarsen the graphite.

(10) Treatment with titanium and carbon dioxide after treatment with calcium silicide prevents the formation of nodules.

(11) Kish graphite may inoculate the melt.

(12) Undercooling may occur.

Preliminary Cooling-Curve Experiments and Effect of Cooling Rate

It was felt that in the preliminary experiments on nickel–carbon alloys very little control had been exercised over cooling rates by casting ingots in green-sand moulds and that, in addition, variations in pouring temperature were certain to have occurred. Therefore the effect of cooling rate was investigated using a different melting technique. In each case 60 g. of pure nickel shot were melted in a graphite crucible machined from unused electrode carbon. The internal dimensions of each crucible were approximately 1 in. dia. by $3\frac{1}{4}$ in. deep, and the wall thickness was about $\frac{1}{4}$ in. A thermocouple in a thin silica sheath was placed through a hole in the graphite lid so that the hot junction was at the centre of the ingot. The thermocouples for all the work reported here were Johnson-Matthey platinum/platinum–10% rhodium, 0·25 mm. in dia., and they were sheathed with long sealed-end translucent silica tubing 0·125 in. in dia. The crucible was placed in each case at the centre of a platinum-wound tube furnace $2\frac{1}{8}$ in. in bore and 20 in. long, and was supported on a graphite stand, from which it was insulated by a $\frac{1}{4}$-in. thick silica disc. The charge absorbed carbon from the crucible walls to the maximum amount soluble at the highest temperature reached, that is, 1500° C. in each case. From this treatment every ingot became hypereutectic.

To investigate the effect of cooling rate, four ingots were made ; these were cooled as follows :

Ingot No.	Cooling
15	From 1500 to 1100° C., at rate of 100° C./20 min.
16	From 1500 to 1100° C., at rate of 100° C./12 min.
17	From 1500 to 1100° C., at rate of 100° C./8 min.
18	Taken from furnace at 1500° C. and air-cooled.

Ingots 15, 16, and 17 all had similar structures. The central zones in each case were entirely devoid of kish graphite, appearing exactly

eutectic in composition ; a typical spot is illustrated in Fig. 53. With decreasing cooling rate the graphite in these zones becomes slightly coarser. At the tops of the ingots segregations of kish graphite occurred ; a typical area is illustrated in Fig. 52. The amount of this segregation decreased with increasing rate of cooling. The flake graphite also became slightly coarser near the crucible walls and in the vicinity of the silica sheath of the thermocouple ; Fig. 54 shows the coarser graphite near the silica sheath of ingot 16. These melts are apparently susceptible to inoculation by graphite and to coarsening in the presence of silica. The latter point will be referred to later in more detail.

Ingot 18 had a structure completely different from the others cooled more slowly. At the centre of the ingot was a small area of undercooled graphite with primary metal dendrites and some distorted hypereutectic graphite ; this area is shown in Fig. 55. The remainder of the ingot had nodular graphite in the metallic matrix, with occasional distorted hypereutectic graphite flakes (Fig. 56). At a high magnification the nodules were found to be rather irregularly formed spherulites ; a typical example is shown in Fig. 57. Apart from the irregular outline, these graphite spherulites are exactly similar to those found in malleable iron of high sulphur content. We have now, therefore, two methods of producing nodular graphite structures in nickel–carbon alloys, *viz.*, (*a*) by treatment with calcium silicide or (*b*) by rapid cooling.

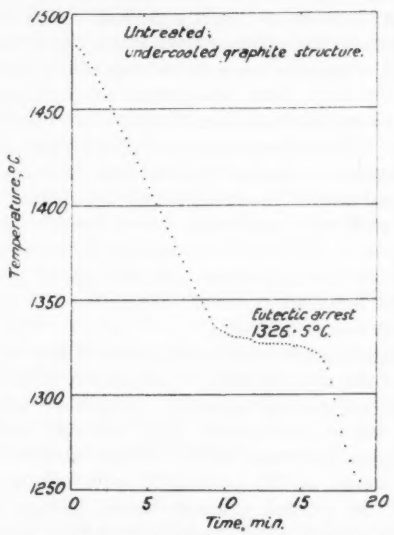

FIG. 127—Cooling curve of hypereutectic nickel–carbon alloy melted in a graphite crucible. Ingot 19. Superheated to 1500° C. No addition

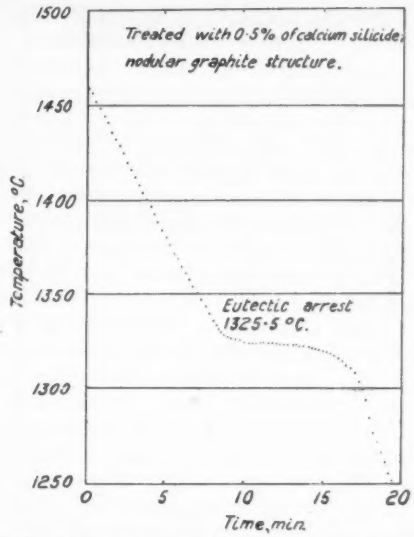

FIG. 128—Cooling curve of hypereutectic nickel–carbon alloy melted in a graphite crucible. Ingot 20. Superheated to 1500° C. and 0·5% of calcium silicide added to give a nodular graphite structure. (*See* Figs. 59 and 63)

Cooling curves were then taken on three further ingots melted in a manner similar to that described for ingots 15 to 18, except that in the case of ingots 20 and 21 a graphite tube was fitted to the top of the crucible leading out of the furnace. This facilitated the making of the additions. The temperature readings were taken on a null deflection potentiometer to 5 microvolts. The ingots were treated as follows :

Ingot No.	Treatment.
19	Superheated to 1500° C. No addition.
20	Superheated to 1500° C. Calcium silicide (0·5%) added.
21	Superheated to 1500° C. Sulphur (0·3%), as roll sulphur, added.

The additions were stirred in with a fine silica tube. The cooling curves obtained are given in Figs. 127, 128, and 129.

Many other cooling curves have been taken on nickel–carbon alloys in the untreated condition and after treatment with calcium silicide and sulphur. In all these the eutectic arrest for undercooled graphite and for spherulitic structures occurred at almost identical temperatures—the actual values all lay between 1323° and 1325° C. In the case of the coarse graphite structures produced by the addition of sulphur, the eutectic arrest occurred at temperatures ranging between 1310 and 1330° C., the actual temperature depending upon the sulphur content. No

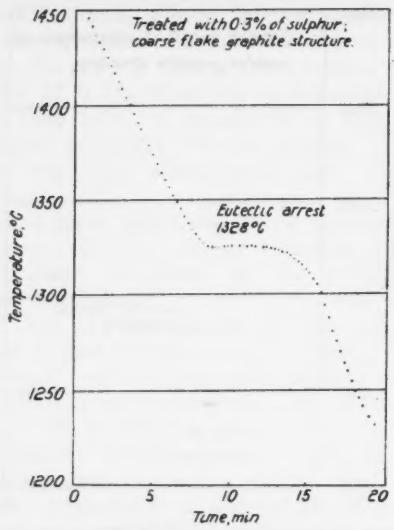

FIG. 129—Cooling curve of hypereutectic nickel-carbon alloy melted in a graphite crucible. Ingot 21. Superheated to 1500° C. and 0·3% of roll sulphur added to give coarse flake graphite. (*See* Figs. 60 to 62)

information was available upon the nickel–carbon–sulphur system, but the nickel–sulphur system[47] shows a sharp drop in the liquidus temperature with increasing sulphur contents until a eutectic of nickel and nickel sulphide occurs at 645° C. and 21·5% of sulphur. From this it might be expected that the addition of sulphur to nickel–carbon alloys would cause a drop in the temperature of the eutectic arrest. If this is true, it would explain the coarse graphite eutectic solidifying at various temperatures which may be lower than in those of untreated alloys with undercooled structures. In the case of the cobalt–carbon alloys subsequently to be described, it was possible to show that a coarse flake graphite alloy had a much higher eutectic arrest than an alloy of the same sulphur content with undercooled graphite. No method was discovered to produce coarse flake graphite in nickel–carbon alloys without the addition of either sulphur or tellurium, and it is possible that the latter element has a similar effect to that of sulphur in lowering the eutectic arrest temperature, irrespective of the type of graphite structure.

The microstructure obtained in ingot 19 was very similar to that described and illustrated for ingots 15 to 17. Ingot 20 had well-developed spherulites over most of its length* (Fig. 58), except near the top, where a very heavy segregation of spherulites occurred (Fig. 59). It would seem that these spherulites represent the hypereutectic graphite. If so, the nodules must have formed either directly from the melt or from another solid phase after the eutectic had solidified. Only by such means could the upward flotation be explained. The structure of a well-formed spherulite at a high magnification is shown in Fig. 63. The structure of ingot 21, treated with sulphur, had uniformly coarse graphite (Fig. 62), becoming very coarse and in greater amount at the top of the ingot. The graphite in this sample, being coarse, was easy to examine for internal structure, and under polarized light the flakes were found to have similar structures to those found in the graphite flakes of cast irons and reported elsewhere.[10] Typical examples of the internal graphite structures are given in Figs. 60 and 61.

* For micro-examination, complete vertical sections were taken of each ingot.

Occurrence of Nodular Structures in "As-Cast" Cast Irons.

Thus far it had been possible to show that nickel–carbon alloys possess many structural features in common with cast irons, but they appeared at first to differ in one unique and very important aspect, *viz.*, that nodular graphite structures of the spherulite type could be produced easily whenever desired in as-cast materials without any heat-treatment. In cast irons, nodular structures could be obtained only by the heat-treatment of a white cast iron at an elevated temperature for a lengthy period. It is of great commercial importance that when the graphite is in this form the cast irons possess considerable ductility and shock resistance, whereas with the graphite in the flake form they are non-ductile and brittle. If nodular structures of graphite could be produced in cast irons without heat-treatment, the material would assume, by virtue of its low cost and enhanced properties, much greater competitive value. With these considerations in mind it was decided to examine a wide range of cast irons to find whether nodular structures did occur in the as-cast condition.

Several instances of nodular structures in as-cast cast irons were found, some apparently forming by different mechanisms. Nodular structures were found associated with inverse chill; the following instance will suffice: Reference has been made to the ease with which inverse chill occurs in refined graphite irons having a completely undercooled graphite structure, and to the fact that complete refinement of the graphite can be achieved by additions of ferrosilicon-

zirconium. A charge of Swedish white iron and steel scrap was melted in a coke-fired forced-draught crucible furnace; when molten, ferro-silicon-zirconium was added to give material of the following analysis after pouring into 1·2-in. test-bar moulds in green sand:

Total carbon... ...	3·28%
Silicon	1·19%
Manganese	0·53%
Sulphur	0·023%
Phosphorus	0·030%
Zirconium	0·547%

The general structures of these bars is shown in Fig. 64 and consisted of fine undercooled graphite in a matrix of pearlite. The usual complex zirconium sulphide and zirconium carbide particles were present in this material. An area of inverse chill is illustrated in Fig. 65. Nodules of graphite are seen in transformed austenite dendrites with the interdendritic regions containing a laminated or " platey " white-iron eutectic. Fig. 66 shows the structure of this white-iron eutectic at a high magnification, where its pattern is clearly registered. In the eutectic of a normal white-iron structure the austenite or transformed austenite is the dispersed phase and the cementite is the continuous phase, but in this case the cementite phase shows little approach to continuity. Fig. 67 shows a nodule at a higher magnification. The peculiar nature of the white-iron eutectic in these illustrations should be noted and compared with that usually occurring in white irons (many examples are shown in the literature); reference will be made to this in a later section. Without doubt here we have a clear occurrence of spherulitic graphite nodules in cast iron which has not been subjected to heat-treatment.

A further case of the occurrence of nodules was found in a pair of cast-iron scissors of very thin section ($\frac{1}{8}$ in.); the micrograph of the structure (Fig. 68) shows a ferrite matrix with a network of phosphide eutectic and pearlite. The composition of the iron was:

T.C.	Si	Mn	S	P
3·3%	3·4%	0·66%	0·120%	1·43%

Taking into account both silicon and phosphorus, it will be observed that the iron is hypereutectic. Hanemann and Schrader [76] have illustrated a case of the occurrence of graphite spherulites in a single-cast piston ring, 80 mm. in dia. and 3 × 3 mm. in section, of the following composition:

T.C.	Si	Mn	S	P
3·75%	2·59%	0·50%	0·088%	0·54%

Again the iron was hypereutectic and cast in a very thin section. Several cases of this type have now been observed in hypereutectic cast irons which have been rapidly cooled. Occasionally the spherulites, which are usually very small, have flake graphite attached, as shown in Fig. 69. At this point it is convenient to re-examine the structure illustrated in Fig. 25. This, it will be recalled, shows the central zones of an ingot melted *in vacuo*, using a graphite crucible. The hypereutectic graphite has floated to the top of the ingot and in the central zones undercooling has occurred, giving transformed austenite dendrites and undercooled graphite. Fine nodular graphite is also to be seen in this micrograph. Nipper [8] has shown a similar case of spherulitic graphite in a vacuum-melted hypereutectic iron, and Gröber and Hanemann [1] have illustrated graphite spherulites in a sample containing 7·7% of carbon, quenched from 2250° to 160° C. It would thus appear that spherulitic carbon nodules can be produced by fairly rapid cooling of hypereutectic melts. At this stage no evidence was available to indicate whether these nodules formed directly from the melt or by the decomposition of a carbide after solidification. In some of the cases quoted, undercooling had resulted from rapid cooling, but in other cases undercooling was produced by vacuum melting, and the cooling rates were slow. To produce spherulites of this type, therefore, it seems that severe undercooling of a hypereutectic cast iron must occur. It is conceivable that these nodules represent the hypereutectic constituent, which may be graphite or carbide, deposited down the supersolubility curve of the hypereutectic side of the constitutional diagram (Fig. 7*c*). This would not, however, explain the presence of primary austenite dendrites. A further explanation would seem to be that undercooling follows the extension of the normal hypereutectic liquidus until the hypo-eutectic side of the supersolubility curve is reached. The graphite deposited over this temperature range would correspond to the spherulitic variety. The melt would then deposit primary dendrites of austenite until the hypertectic point is reached, when a binary complex of austenite and cementite solidifies and subsequently decomposes to give undercooled graphite (Fig. 7*d*). However, in the authors' opinion, a more likely explanation would be that solidification proceeds along the normal hypereutectic liquidus, depositing normal kish graphite, until the eutectic point is reached. From this point solidification then follows the undercooled extension of the normal hypo-eutectic liquid giving austenite dendrites. Further solidification follows the supersolubility curve on the hypereutectic side until the hypertectic point is reached. The

spherulitic graphite originates from the phase deposited down this portion of the supersolubility curve (Fig. 7*e*).

If these spherulites form as a result of the decomposition of cementite deposited down the hypereutectic supersolubility curve then we have an interesting extension of the findings of Eash that the undercooled eutectic solidifies first as a white-iron eutectic. The mechanism of spherulite formation in nickel–carbon alloys was unknown, and spherulite formation in heat-treated malleable iron was the only case where definite knowledge of the process existed. Could spherulites be formed directly from the melt ? A search was therefore made through the literature to find whether any general rules existed about undercooling and the formation of spherulites. The authors were impressed by the general lack of information on spherulitic crystallization. References 77 to 91 were among the most important examined. The undercooling of a few organic- and inorganic-salt eutectics has been studied in some detail, but little or no reference is made to the type of crystallization produced by different degrees of undercooling. Spherulitic formations are well known to the mineralogist, but practically no information is available from this source upon how they form. One investigator briefly refers to the production of spherulites by the solidification of liquid crystals and also by the rapid cooling of a molten organic compound into the region of labile crystallization. This latter point is particularly interesting in view of the fact that rapid cooling of nickel–carbon alloys gives graphite spherulites. Two American investigators have produced artificial spherulites and related aggregates by allowing a concentrated solution of one salt to diffuse into a gel containing a dilute solution of another salt able to react with the first by metathesis. It should be noted that these conditions do distantly resemble those obtaining when spherulites are formed by heat-treatment of a white iron (containing iron sulphide). These investigators point out that spherulites must grow by the simultaneous crystallization of fibres radiating from a common centre in all directions with equal speed. Such a mechanism would lead to a spherical shape at all times during their growth. However, during the course of their experimental work and in agreement with certain German investigators, they advance the hypothesis that spherulitic crystallization may also begin from a centre within two opposite solid angles, together less than 360°, and that at a further stage the outward growth of the fibres ceases and further crystallization takes place by filling-in the waist. No evidence of this mode of spherulitic formation has been found in any of the materials mentioned in this report.

Since it was well known that spherulites could be produced in cast iron by the decomposition of cementite during heat-treatment of a white iron at 650–1050° C., and as the possibility of producing graphite in this form directly from the liquid seemed too nebulous at this stage, an attempt was made — avoiding heat-treatment — to obtain spherulites, basing the procedure on known facts. The following gives the train of reasoning leading up to the experimental work in this context.

It was assumed that a mechanism was required which would enable an iron to solidify white and subsequently, after complete solidification, to graphitize without heat-treatment. This process would be essentially the same as that employed in the malleable-iron process. The only situation known definitely to lead to suitable conditions would have to be based upon the reaction between manganese and sulphur in cast iron. In the absence of manganese the sulphur exists as iron sulphide, but manganese (if present), having a greater affinity for sulphur, replaces it with manganese sulphide if added in sufficient excess. Iron sulphide causes the stabilization of iron carbide, whilst manganese sulphide behaves neutrally in this respect. The reaction between iron sulphide and manganese can be expressed as follows :

$$FeS + Mn \rightleftharpoons MnS + Fe.$$

This reaction is reversible, tending to proceed to the left-hand side at high temperatures and to the right-hand side at lower temperatures. Owing to this reversibility it is necessary, if no iron sulphide is required, to have manganese present in excess of that required theoretically to give manganese sulphide according to the formula MnS. Even when at room temperature only manganese sulphide is present, appreciable amounts of iron sulphide may be formed at elevated temperatures (1000–1600° C.), owing to the reversibility of the reaction. It was thought, bearing this in mind, that by adjusting the sulphur and manganese contents so that during solidification a sufficient amount of iron sulphide was present to cause white-iron solidification, and that if after solidification—consequent upon decreasing temperature—sufficient sulphur became converted to manganese sulphide, the white-iron structure would be enabled to graphitize automatically. Trials of a wide range of manganese and sulphur contents were made, but it seemed that for the principle to work the presence of an additional graphitizing element was required. With irons of the composition indicated below it was found possible to produce a form of graphite which

probably formed by the decomposition of cementite after solidification :

T.C., %	S, %	Mn, %	S, %	Ni, %
2·8–3·2	0·7–1·3	0·1–0·3	0·2%–0·6	1·5–7·0*

* Or 1–6% of copper.

The type of graphitization produced by such compositions has been described and illustrated in another paper,[92] dealing with the neutralization of sulphur by various alloying elements. It is generally characterized by a mesh-like pattern with no individual graphite flakes, all the graphite in one eutectic cell being interconnected. From the arrangement of some of this graphite it appears to outline what were originally islands of cementite and, occasionally, where graphitization is not complete, it can actually be seen rimming areas of eutectic carbide. On rare occasions a few very fine spherulites were observed. These have also been illustrated elsewhere.

It was felt that these findings justified the assumption that the manganese–sulphur reaction could be used to produce graphitization after solidification. This process could under certain conditions proceed to completion but, except on a very small scale, only a few isolated spherulites were produced. It was therefore decided that it would be necessary to provide a number of nuclei at which spherulitic crystallization could begin. After many experiments the following procedure was found to be slightly more successful in producing spherulites. Melts were made with compositions within the following range :

Total carbon... ...	2·6–3·2%
Silicon	1·8–2·6%
Manganese	0·1–0·3%
Sulphur	0·01–0·1%
Phosphorus	0·03–0·1%

When molten, the melts were treated with 1–3% of calcium silicide and then with about 0·1–0·4% of sulphur. It was argued that the calcium silicide would provide the nuclei and then the sulphur addition would provide the required manganese–sulphur reaction. No irons were produced by this method with all the graphite in the form of spherulites, although these were much larger and more numerous than previously ; typical examples are shown in Fig. 70. Occasionally the graphite assumed a pseudo-Widmanstätten pattern (Figs. 71 and 72), and sometimes occurred in a very fine spherulitic pattern attached to flakes of graphite (Fig. 73).

We have now described three distinctly different occurrences of spherulitic graphite formations in as-cast cast irons, *viz.* :

(1) Associated with inverse chill in irons also containing undercooled graphite.

(2) In undercooled hypereutectic grey irons.

(3) In irons of low manganese and high sulphur contents.

With this knowledge there appeared to be justification for proceeding with the investigation of nickel–carbon spherulitic formations, but, before describing this, one other group of experiments must be referred to. Remembering that additions of calcium silicide to nickel–carbon alloys did not produce spherulites in the presence of sulphur, it was thought that if this element could be eliminated in cast irons, spherulites might also be produced by calcium silicide. Synthetic melts were made with sulphur contents as low as 0·001%, but all these reacted in the normal way with calcium silicide, giving flake graphite.

The Effect of Calcium Silicide on Nickel–Iron–Carbon Alloys

As calcium silicide did not produce graphite spherulites even in cast irons of very low sulphur content, it was felt that the production of spherulites by this method may be a property of nickel–carbon alloys alone. It was therefore decided to investigate the effect of iron upon this tendency in a series of hypo-eutectic nickel–iron–carbon alloys ranging from the nickel–carbon to the iron–carbon side of the ternary system. The ternary iron–nickel–carbon system has been investigated by Kase[70] and his results have been redrawn and criticized by Marsh,[93] who has also included those obtained by Schichtel and Piwowarsky.[94] Many points are still in doubt, but with safety it can be said that the liquidus surface shows a trough extending from the binary eutectic on the iron–carbon to the binary eutectic on the nickel–carbon side. Fig. 130, taken from Marsh, indicates the effect of nickel on the carbon content and temperature of the eutectic.

Eleven melts were carried out in a high-frequency induction furnace, using charges made up of pure nickel, electrode carbon, and Lowmoor iron. A representative analysis of the iron was :

C	Si	Mn	S	P
0·18%	0·02%	0·14%	0·004%	0·026%

After superheating to about 1500° C., one ingot 6 in. long by ½ in. dia. was poured in green sand from every melt. About 0·8% of calcium silicide was then added and a further similar ingot was cast. Samples for chemical analysis and microscopic examination were taken from the bottom half of each ingot. The analyses and microstructures of these ingots are summarized in Table II. All the analyses were carried out on the calcium silicide treated bars.

Melts 22 to 29 were carried out first and then followed by melts 30 to 32, to scrutinize the

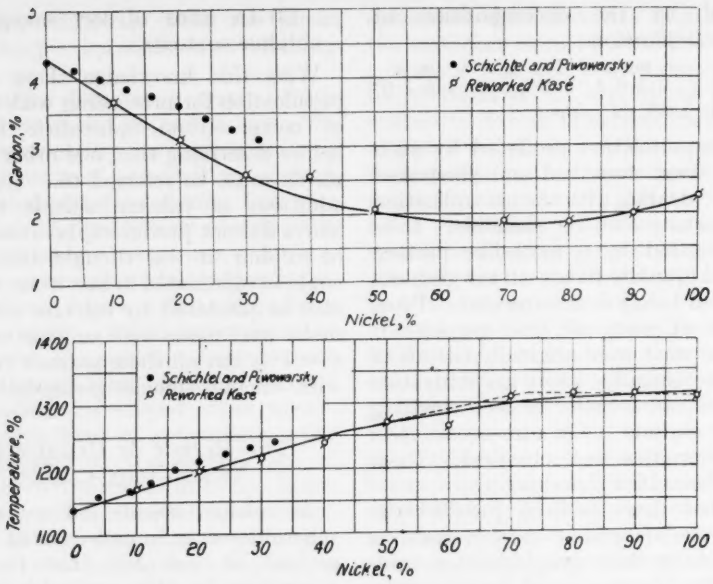

FIG. 130—Effect of nickel on the temperature and carbon content of the eutectic in iron–nickel–carbon alloys (Marsh)

TABLE II—*Effect of Calcium Silicide Additions to Nickel–Iron–Carbon Alloys*

Melt No.	Fe, %	Ni, %	C, %	S, %	Si, %	Treatment	Microstructure
22	5·74	...	1·60	...	...	Nil	Undercooled
						CaSi	Nodular
23	9·84	...	1·50	...	...	Nil	Undercooled
						CaSi	Nodular
24	23·8	...	1·72	...	...	Nil	Undercooled
						CaSi	Nodular
25	40·5	...	1·57	...	...	Nil	Undercooled
						CaSi	Nodular
26	57·86	...	1·57	...	...	Nil	Undercooled
						CaSi	Flake
27	74·33	...	2·22	...	...	Nil	Undercooled
						CaSi	Flake
28	85·79	...	2·41	...	...	Nil	Undercooled
						CaSi	Flake
29	90·81	...	2·87	...	...	Nil	Undercooled
						CaSi	Flake
30	46·83	50·72	1·58	0·01	0·89	Nil	Undercooled
						CaSi	Nodular
31	48·79	48·06	2·08	0·01	1·00	Nil	Undercooled
						CaSi	Nodular
32	53·0	44·61	1·93	0·01	0·61	Nil	Undercooled
						CaSi	Flake

region 45–55% of iron in greater detail. Inspection of the results clearly reveals a change in the reaction of these alloys to additions of calcium silicide at some iron content lying between 49 and 53%. Typical microstructures of this series are represented by Figs. 75 and 76, taken from melt 31, and Figs. 77 and 78, taken from melt 32. From these results the present authors concluded that iron in amounts greater than about 50% confers, either directly or indirectly, upon nickel–iron–carbon alloys a property by virtue of which they are susceptible to nucleation by calcium silicide to give flake graphite. For iron contents of less than about 50%, calcium silicide behaves in a different manner and certainly does not inoculate the melt, spherulitic crystallization being produced.

Formation of Undercooled Graphite in Cast Irons

It will be recalled that reference was made to the work of Eash, who found that irons giving undercooled graphite structures solidify white and subsequently decompose to give undercooled graphite after solidification. It was also pointed out that this investigator's methods of demonstrating this might be considered somewhat crude and the composition of his irons unusual. In view of the importance of Eash's suggestions, the present investigators decided to try to confirm them, using different experimental methods. It should be stated immediately that the results which were eventually obtained entirely confirmed the claims made by Eash.

For these experiments two irons were used, one a low-carbon iron (in this respect similar to that used by Eash) and the other one of a more normal composition. The analyses of these two irons are as follows :

	Low-Carbon Iron	Normal-Carbon Iron
Total carbon, % ...	1·92	3·10
Silicon, % ...	2·68	2·05
Manganese, % ...	0·74	0·90
Sulphur, % ...	0·036	0·045
Phosphorus, % ...	0·03	0·04
Titanium, % ...	0·16	0·21
Vanadium, % ...	0·27	0·31

In contrast to the iron used by Eash, these irons contained titanium (and vanadium) in sufficient amounts to guarantee the formation of undercooled graphite structures, even with relatively slow rates of cooling.

Small ingots of these irons weighing about 40–50 g. were melted in fireclay crucibles in a platinum-resistance tube furnace placed vertically above a bath of water about 2 ft. deep. By moving a lever the bottom of the furnace could be instantly disengaged and the contents of the furnace, including the ingot in the crucible, plunged into the bath of water. Temperatures were measured using platinum/platinum–10% rhodium thermocouples and a null-deflection quick-reading potentiometer, readings being taken to 5 microvolts. Cooling curves were taken by recording the temperature every 15 sec.

Melts of the high-carbon material were made first. It was found that ingots slowly cooled to room temperature in the furnace gave an undercooled graphite structure such as that illustrated in Fig. 79 of the ingot from melt 33. On many occasions ingots of 40–50-g. weight were quenched within 5° C. below the eutectic arrest. In no case was a completely white structure produced, but in many a partially white ingot was obtained. Fig. 131 shows the cooling curve of ingot 34, which was rapidly quenched immediately after solidification. Fig. 80 shows a vertical section through this ingot, and Fig. 81 shows the white-iron structure at a higher magnification. The grey portions of the melt had undercooled graphite in a matrix of martensite. Similar experiences were obtained with the low-carbon material ; by quenching immediately after solidification no completely white samples were obtained.

Whilst the delay between the taking of the last temperature reading and plunging the ingot into the quenching bath was only a fraction of a second, it was felt that either the effectiveness of the quench was inadequate, owing to the crucible surrounding the ingot, or that the breakdown of

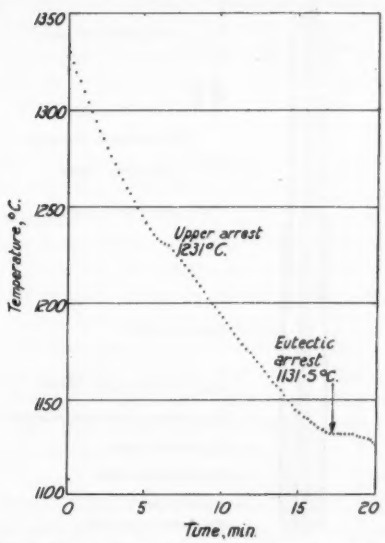

Fig. 131—Cooling curve of 50-g. cast-iron ingot (ingot 34) quenched immediately after solidification. (*See* Figs. 80 and 81)

the undercooled white iron was very much more rapid than was indicated by the work of Eash. It was further possible that the cooling curves were not indicating complete solidification. Accordingly, it was decided to use very much smaller ingots, *viz.*, of from 5 to 7 g. These were in crucibles made from silica tubing, and the arrangement, together with the approximate dimensions, is illustrated in Fig. 132. Using this arrangement ingots were obtained which were almost completely white when quenched just after solidification. A typical cooling curve for a melt of the higher-carbon material, using a 6-g. ingot, is shown in Fig. 133, and the structure of a section of the quenched ingot is seen in Fig. 84. The duration of the arrest in this case was as long as that of similar ingots which were allowed to cool to room temperature at the same initial rate, but which were completely grey. The very small grey portions visible in this ingot could not have been responsible for the arrest.

From these results it was apparent that the breakdown of this carbide in undercooled samples is extremely rapid and may even be proceeding before the whole of the eutectic is solid. It should be noted that the eutectic found in these quenched samples (Figs. 81 and 84) is very similar to that found in the inverse chill of fine undercooled graphite irons (Figs. 65 and 66). The same plate-like characteristics are visible. That this structure differs from that of a normal white iron is shown by Fig. 82, which illustrates the general structure of a 50-g. ingot of the higher-carbon material to which an addition of 2% of chromium had been made and which was cooled at the same rate, during solidification, as the ingot illustrated in Fig. 81. This may provide a clue to the formation of inverse chill in undercooled irons, since the white iron represents the eutectic, which, in that particular locality, instead of decomposing after solidification to give undercooled graphite, remains stable to room temperature. Occasionally the carbide of the inverse chill does begin to graphitize at a lower temperature than is required to give undercooled graphite, giving nodular structures of the type shown in Figs. 65, 66, and 67. Why the white-iron structure at the centre of the section should show this increased persistence is not apparent, but it should be remembered that it shows other features of metastability differing from normal white iron, for instance, its apparent immunity to graphitizing elements, etc.

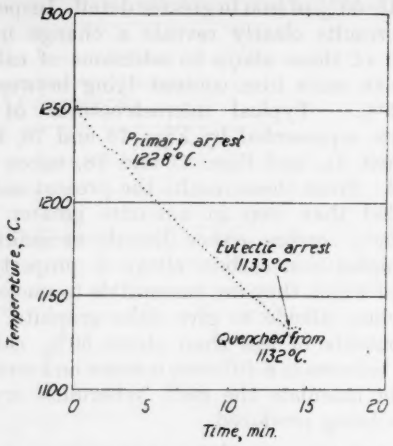

FIG. 133—Typical cooling curve obtained with a 6-g. cast-iron ingot melted in the arrangement shown in Fig. 132. (For structure *see* Fig. 84)

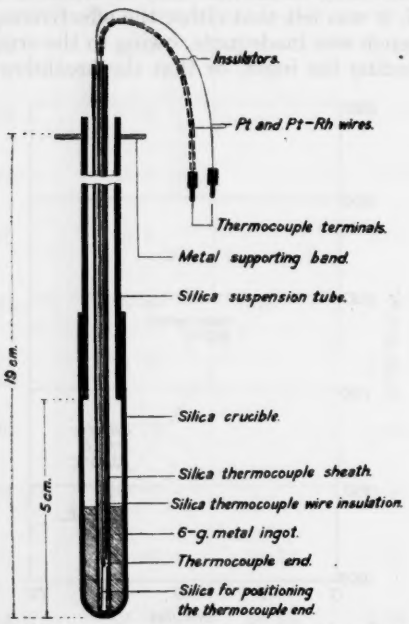

FIG. 132—Arrangement of a silica crucible and thermocouple for quenching experiments with 5–7-g. ingots

FORMATION OF UNDERCOOLED GRAPHITE IN NICKEL–CARBON ALLOYS

As undercooled graphite structures, as well as flake graphite, could be produced in nickel–carbon alloys, it was decided to investigate whether in these alloys also an unstable eutectic solidified first and then decomposed to give undercooled graphite. Initially the experiments were carried out on 50-g. ingots of hypo-eutectic melts in fireclay crucibles, but, as with the cast irons, no

positive results were obtained. Next, 6-g. ingots were melted in silica crucibles similar to those used for cast irons, but the nickel–carbon melt reacted with the silica with a marked evolution of gas, which invariably lifted the melt to the top of the crucible and even blew the molten metal up the very narrow space between the thermocouple sheath and the silica positioning tube. This probably reflects an important difference between cast irons and nickel–carbon alloys, the carbon in the latter being much more reactive to oxides and oxygen. This has been referred to previously in connection with the remelting and desulphurizing with Na_2CO_3 of nickel–carbon alloys. In the particular case referred to here, the following reaction probably occurs :

$$2C\ (Ni) + SiO_2 = Si + 2CO$$

The experiments were therefore continued using 6-g. ingots melted in graphite crucibles of the same size as the silica crucibles used previously. This method proved satisfactory. Sharp arrests were obtained, and, owing to the high thermal conductivity of graphite, more efficient quenching resulted. Ingots were heated to 1400–1500° C. and cooled in the furnace. Those cooled to room temperature or quenched more than 50° C. below the eutectic arrest showed undercooled graphite with some kish graphite, particularly at the tops of the ingots. The crucible walls had the usual faint coarsening effect on the graphite. The ingots cooled in the same manner but quenched immediately after solidification had structures consisting almost completely of spherulitic graphite with, adjacent to the crucible wall, a thin band of flake graphite followed by a still thinner band of very fine undercooled graphite. Fig. 134 shows a cooling curve of a 6-g. nickel–carbon ingot up to the point of quenching, and Fig. 83 illustrates the general and edge structure

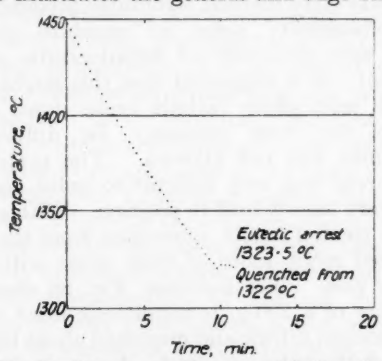

FIG. 134—Cooling curve of a 6-g. nickel–carbon ingot melted in a graphite crucible and quenched immediately after solidification. (For structure, *see* Fig. 83)

of the same ingot. The coarse graphite produced by the crucible wall is clearly seen. The graphite nodules in this sample were spherulites exactly similar to those illustrated previously in this paper, but of much smaller size. It would appear from this group of experiments that undercooled graphite structures in nickel–carbon alloys arise as the result of the decomposition of an unstable eutectic, probably nickel solid solution + nickel carbide (Ni_3C), shortly after solidification. This carbide may be very much more unstable than that formed under similar conditions in cast irons and so decomposes at a somewhat lower temperature during the quench to give the spherulitic graphite which is seen in Fig. 83. Near the crucible wall the graphite crucible has inoculated the melt to prevent undercooling, and this flake graphite has in turn rendered the undercooled carbide bordering it more unstable than that some distance away so that it decomposed at its normal temperature, giving characteristic undercooled graphite.

Now we have a clue to the formation of nodular structures in nickel–carbon alloys by rapid cooling (referred to previously, *see* Figs. 56 and 57). In cast irons which normally would give undercooled graphite structures, an increase in the rate of cooling gives progressively finer graphite until, in the limiting case, the cooling rate is so rapid that the white eutectic does not have time to transform and a white iron results. In nickel–carbon alloys a similar state of affairs exists, but although the very rapid cooling does not allow time for carbide decomposition at a high temperature, it cannot prevent it from decomposing at a lower temperature, owing to the extreme instability of the carbide. At lower temperatures spherulites are produced.

Formation of Spherulites in Nickel–Carbon Alloys by Calcium Silicide Additions

Examples of the effect of calcium silicide have already been shown and discussed, but the findings just described suggest an explanation of the calcium silicide effect in nickel–carbon and nickel-rich nickel–iron–carbon alloys. It will be recalled that the freezing point of the eutectic in calcium silicide treated ingots was very near to that of untreated ingots giving undercooled graphite. This has been confirmed several times, the temperature of the eutectic arrest in both cases being 1325° C. ± 1. It is believed that this temperature represents the solidification of the eutectic of nickel carbide and a solid solution of carbon in nickel. The function of calcium silicide is to stabilize the carbide over a greater range of

temperature, or to provide the right kind of nuclei for spherulitic crystallization, or a combination of these. The effect of the first possibility would be to cause carbide decomposition at a lower temperature than is normal. The fact that rapid cooling also produces spherulites in nickel–carbon alloys would seem to indicate that their production by the addition of calcium silicide is not necessarily a process involving the introduction of nuclei into the melt. It might be argued that cast irons differ in this respect from nickel–carbon alloys. This is true in so far as calcium silicide does not promote the stability of carbides in iron-rich alloys, but it is possible that the undercooled white iron, normally parent to undercooled graphite, may decompose at lower temperatures to give nodular structures ; this has already been quoted in the section preceding the previous one, as an explanation of spherulites occurring with inverse chill (*see* Figs. 65, 66, and 67).

If this is correct we now have the problem of producing spherulitic graphite in a cast iron without heat-treatment stated in a further form. The white-iron structure giving rise to undercooled graphite in cast irons already fulfils some of the conditions for nodule formation—it solidifies white and after solidification it is very unstable, changing to austenite and graphite. If this carbide can be stabilized to lower temperatures, or if the appropriate nuclei for spherulitic formation can be introduced, the problem will be solved.

So far in this discussion we have referred only to the spherulites produced in nickel–carbon alloys which originate from the eutectic portion of the alloy, but it has already been shown (Figs. 35 and 59) that hypereutectic graphite can also be obtained in this form by treatment with calcium silicide. Furthermore, very rapid cooling of hypereutectic melts will also give nodules. Thus if the explanations proffered so far are correct for the eutectic, they must also apply to the hypereutectic constituent. Now, in Figs. 58 and 59 it has been shown that the hypereutectic spherulites segregate at the tops of the ingots as if by flotation. It is suggested here that these most probably floated to that position—at least partially—as graphite, since nickel carbide would not be expected to have such a low relative density as to permit such a marked flotation. Therefore the hypereutectic graphite spherulites must have formed as the result of the decomposition of hypereutectic carbide suspended in the liquid. The process by which this occurs is not easy to imagine, because in the case illustrated the nodules are quite large and would therefore begin to rise in the melt before reaching their final dimensions and continue to grow on their journey upwards. For the graphitization of a hypereutectic nickel carbide during solidification, we have perhaps :

$$Ni_3C \longrightarrow 3Ni + C,$$

but in the case under consideration the reaction would take place in liquid surroundings, and the nickel product of the reaction would hold more than the eutectic concentration of carbon in solution and, furthermore, it would be liquid, since—according to the equilibrium diagram and the above considerations—the only phases which can be present in hypereutectic alloys above the eutectic temperature are liquid, graphite, or nickel carbide. Thus the explanation for the formation of spherulites by calcium silicide additions based upon an increased stability of a carbide becomes more difficult when applied to hypereutectic melts, and the possibility of calcium silicide providing the appropriate nuclei for spherulitic crystallization looms larger. The general literature on the subject of spherulitic formation indicates, without specifically saying so, that one condition for the process is restricted growth. In white cast irons being annealed, one of the present authors[96] has shown that the determining factor is whether the right nuclei (FeS) are present, and not the extent of the restriction of growth, since temperature has no effect on the type of crystallization.

Nickel Carbide in Nickel–Carbon Alloys

In what has been said so far, the existence of a nickel carbide has been tacitly assumed and it has been suggested that its existence in nickel–carbon alloys was not established. The compound had, however, been produced by the reaction of nickel powder and carbon monoxide. By quenching hypereutectic ingots in water directly from the liquid state nodular structures predominated, but occasionally areas of graphite occurred which were obviously of interdendritic pattern (Fig. 85). It is suggested that this might represent an area where carbide broke down *in situ* because the time necessary for diffusion to spherulites was not allowed. The graphite in these areas was very difficult to polish, seeming to be very loosely held in position. With calcium silicide treated ingots, quenching from the liquid produced many more of these areas with much greater ease. For instance, Fig. 86 shows the structure of a 200-g. hypereutectic ingot treated with calcium silicide and quenched about half-way through the eutectic arrest. Again, it should be noted, many well-developed nodules have formed and many dark, apparently interdendritic, areas of graphite can be seen. The increased ease with

which these are formed in calcium silicide treated ingots could be explained by the effect of this addition in increasing the stability of the carbide.

A number of 6-g. ingots were melted in graphite crucibles and held molten for lengthy periods. After quenching, these ingots did apparently show a nickel carbide. For instance, Fig. 87 shows the structure of an untreated ingot heated to 1500° C., slowly cooled over half an hour to 1450° C., and then quenched rapidly in a bath of water. Nodules of graphite are seen, together with a half-tone constituent. At high magnifications these areas can be resolved into an apparent Widmanstätten structure of nickel carbide in solid solution (Fig. 88). Evidence that the dispersed phase is a carbide is given in the proximity of graphite particles ; the phase always decreases in amount near to, and is never found adjoining, graphite (Fig. 89). Confirmatory evidence was obtained when a portion of the ingot was annealed at 700° C. for 4 hr., the dispersed phase in the Widmanstätten pattern being replaced by fine spots of graphite (Figs. 90 and 91).

Marsh [93] has commented upon Kase's [70] assumption of the complete intersolubility of iron and nickel carbide as a very improbable possibility. This criticism is presumably based upon the idea that nickel carbide [72] has a different structure from that of iron carbide.[97,98] In view of the uncertainty which still exists about the structure of iron carbide it is perhaps not wise to go beyond this, but in support of Marsh's contention we have this apparent change in carbide stability and behaviour in the presence of calcium silicide, occurring somewhere in the region of 50% of iron in the nickel–iron–carbon alloys.

Effect of Silicon on Response of Nickel–Carbon Alloys to Calcium Silicide Additions

Four melts of pure nickel, together with 1%, 2%, 5%, and 10% of silicon respectively (as silicon metal), were carried out in graphite crucibles. The melts were slowly cooled in a platinum-wound furnace. This series was then repeated, but 0·5% of calcium silicide was added in each case before cooling. A summary of the microstructures is given in Table III.

Table III—*Effect of Silicon on Nodule Formation in Nickel–Carbon Alloys*

Silicon Added	Microstructure of Untreated Ingots	Microstructure of Ingots Treated with CaSi
1%	Undercooled graphite	Nodular graphite
2%	Undercooled graphite + traces of nodular graphite + a few particles of coarse graphite.	Nodular graphite
5%	Coarse graphite + nickel silicide (Ni_5Si_2) ...	Nodular graphite + nickel silicide
10%	No graphite visible. Solid solution + eutectic with Ni_5Si_2	No graphite visible. Solid solution + eutectic with Ni_5Si_2

Quenching experiments were also carried out on nickel–iron–carbon melts with iron contents of between 40 and 60%. Melts containing more than about 50% of iron had eutectic carbides after quenching immediately on completion of eutectic freezing. Nodular structures were also present in these alloys. Melts containing less than about 50% of iron and correspondingly more nickel gave no carbides when quenched in the same manner. There seems to be a change in carbide stability at about 50% of iron in these alloys, which roughly corresponds with the point at which they react differently to calcium silicide. The structure of an alloy with approximately 60% of iron, quenched immediately after solidification, is seen in Fig. 92, and Fig. 93 shows the carbide phase at a higher magnification. Fig. 94 shows the structure of an alloy with approximately 40% of iron quenched immediately after solidification. No carbides are visible in this case ; only fine graphite nodules and a small amount of interdendritic graphite are to be seen.

In amounts of up to at least 5%, silicon does not appear to influence the tendency of calcium silicide to give nodular graphite when added to nickel–silicon–carbon alloys. The microstructure of the 5% silicon ingot treated with calcium silicide is shown in Fig. 95, in which nodular graphite and nickel silicide in a solid-solution matrix can be seen. The phases introduced by the addition of silicon were similar in appearance to those shown by Iwaze and Okamoto [99] in the nickel–silicon alloys. At 10% of silicon no graphite was visible, and therefore it would appear that silicon either reduces the solubility of graphite or stabilizes the nickel carbide. Ingots containing 5% and 10% of silicon, respectively, were annealed for varying periods at 900° C., but no more graphite appeared nor did the amount of nickel silicide decrease, so that silicon probably reduces the amount of carbon dissolved.

Effect of Copper on Response of Nickel–Carbon Alloys to Calcium Silicide Additions

Nickel–copper–carbon alloys were prepared with up to 20% of copper. In the untreated condition undercooled graphite was obtained in all cases, and nodular graphite was obtained in every case after treatment with calcium silicide.

The Effect of Alloy Additions to Nickel–Carbon Alloys, the Alloy Additions Containing Calcium or other Alkaline-Earth Elements

Since additions of silicon did not have any effect on the graphite structure of nickel–carbon alloys, it was considered likely that the effect of calcium silicide was due to its own peculiar crystal structure functioning as an appropriate nucleus for graphite formation, or to the alloying influence of calcium. To investigate this a calcium–silicon alloy containing more silicon than the calcium silicide used previously,* and various other alloys containing calcium or other alkaline-earth elements,† were used. The compositions of these alloys are given in Table IV. According to the calcium–silicon equilibrium diagram[100] the calcium silicide used in these experiments so far should contain CaSi and $CaSi_2$. The calcium–silicon alloy listed in Table IV contained 80% of silicon, and hence should have consisted of silicon and $CaSi_2$.

Table IV—*Alloys Containing Alkaline-Earth Elements*

Alloy	Composition
Ca–Si	Si, 80%.
Ca–Mg	Ca, approx. 40%; Na, < 2%; Cl, not more than 0·5%; Mg, remainder
Ca–Cu	Ca, approx. 40%; Na, < 1%; Cl, not more than 2%; Cu, remainder
Ca–Zn	Ca, approx. 40%; Na, < 1%; Cl, < 1%; Zn, remainder
Ca–Pb	Ca, approx. 10%; Na, < 0·5%; Cl, < 0·5%; Pb, remainder
Sr–Mg	Sr, approx. 25%; Na, < 2%; Cl, < 2%; Mg, remainder
Ba–Mg	Ba, approx. 35%; Na, < 2%; Cl, < 2%; Mg, remainder
Ba–Pb	Ba, approx. 30%; Na, < 2%; Cl, < 1%; Pb, remainder

Ingots of 60-g. size were melted in graphite crucibles, using pure nickel shot. As previously, the crucibles were machined from electrode carbon. The melts were heated to about 1470° C. and were then treated with 0·5% of one of the alloy additions.

Well-formed spherulitic graphite was obtained in every instance, with the exception of the melt treated with barium–lead, when the graphite was in the form of fine flakes with some irregular nodules which were not spherulitic. This structure is illustrated in Fig. 96. In every ingot treated with an alloy containing magnesium the eutectic structure of $MgNi_2$ was seen.[101] Fig. 97, showing the structure of an ingot after treatment with calcium–magnesium, and Fig. 98, showing the $MgNi_2$ structure at a high magnification, illustrate this. Unless the addition was stirred into the melt, the magnesium-containing phases and the nodular structures occurred only at the top of the ingot, the remainder of the graphite being in the undercooled form. When stirred in, however, the magnesium-containing phases and the nodular structures were both uniformly distributed. This appears to relate the presence of alloyed magnesium with the formation of spherulites. Traces of undercooled graphite were also found in the ingots treated with calcium–zinc. Similarly, additions of calcium carbide have also been found to produce spherulitic graphite in nickel–carbon alloys.

Since these experiments indicated that magnesium alloyed with nickel–carbon alloys was a factor in the production of spherulites, additions of magnesium were made to a number of melts, using magnesium metal, a nickel–magnesium alloy (Mg, 30%), and a copper–magnesium alloy (Mg, 43%).‡ In every case graphite spherulites were produced. It can be concluded that magnesium and calcium most probably function as alloying elements in giving nodular structures. Barium and strontium may behave in a manner similar to calcium and magnesium, but when a positive effect was obtained from alloys containing them magnesium was also present.

Summarizing, it is felt that the essential feature of an addition for the production of spherulites in nickel–carbon alloys is that it shall contain at least either calcium or magnesium in some

* The analysis of the calcium silicide used in all the experiments described so far was : Silicon 62·09%; calcium 25·1%; aluminium 1·39%.

† These alloys were given for the investigation by the Imperial Chemical Industries, Ltd.

‡ Kindly given by Messrs. J. Stone & Co., Ltd.

suitably protected form, so that either element can enter the melt. No evidence has been brought to light to indicate that the crystal structure of the addition has any influence.

Experiments with Cobalt–Carbon Alloys

Much less work has been carried out on the cobalt–carbon alloys than upon the nickel–carbon alloys, but sufficient data have been accumulated to deepen the analogy between iron–carbon–silicon alloys on the one hand and nickel–carbon and cobalt–carbon alloys on the other. No experiments were carried out on the ternary alloys of cobalt and carbon with iron, silicon, and nickel, but these should provide an interesting field for future investigation.

The cobalt–carbon diagram has been studied by several investigators[102–104] whose work shows a fair measure of agreement in respect of the general features of the liquidus. The high-temperature regions in the system are very similar to those in the nickel–carbon system, showing a eutectic at 1300–1315° C., at 2·8–3·0% of carbon. There has been some disagreement as to the actual location of the eutectic point; Boecker[102] put the eutectic temperature at 1300° C., while Hashimoto[104] put it at 1315° C. The eutectic consists of a solid solution of carbon in cobalt and graphite. Cobalt does not appear to form a stable carbide, the only solid phases reported being graphite and solid solution.

Cobalt from several sources was used in these experiments, and some was discarded owing to the presence of impurities, notably sulphur. For the first experiments cobalt briquettes were used, and spectrographic examination of these revealed the following impurities, in order of amounts present: Silicon, manganese, nickel, magnesium, and iron. Chemical analysis gave 0·290% of sulphur in these briquettes.

By air-cooling such ingots the eutectic graphite was obtained in the fine undercooled form, the general structure showing primary graphite and primary solid solution typical of undercooled alloys. The addition of 0·3% of sulphur to air-cooled ingots gave fairly coarse flake graphite. Again, as in the iron–carbon–silicon and nickel–carbon alloys, we have the coarsening influence of sulphur.

Hypo-eutectic melts made up with the above material and electrode carbon, melted in fireclay crucibles and then air-cooled, gave mixed structures of flake and undercooled graphite such as are shown in Fig. 99. Examination of these alloys at a high magnification revealed the presence of cobalt sulphide (Fig. 100). A purer form of cobalt was then obtained in powdered

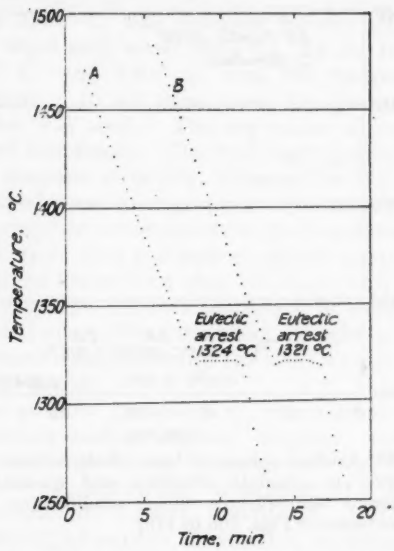

Fig. 135—Cooling curves of two 60-g. cobalt–carbon ingots melted in graphite crucibles: *A*, treated with calcium silicide; *B*, untreated. (For structures *see* Figs. 101 and 102)

form. Chemical analysis showed the following impurities:

Ni	Mn	S	CaO	Al_2O_3	C
0·05%	<0·10%	0·033%	0·99%	0·045%	0·017%

Fe	Cu	P	MgO	SiO_2
0·22%	0·15%	<0·01%	0·24%	0·20%

Two 60-g. melts were carried out using this material melted in graphite crucibles, and cooling curves were taken. The melts were heated to 1500° C. before cooling, one being treated with 0·5% of calcium silicide. The cooling curves obtained are given in Fig. 135. In the untreated sample the eutectic arrest occurred at 1321° C. and in the calcium silicide treated ingot it occurred at 1324° C. The central zones of the untreated ingot had fairly coarse graphite (Fig. 101), the calcium silicide treated ingot had coarser graphite (Fig. 102), and both ingots showed flotation of kish graphite to the tops of the ingots. The general graphite structure became coarser in the vicinity of kish graphite in both samples.

Hypo-eutectic melts using the purer material, carried out in the same manner but air-cooled, gave mixed structures of very fine undercooled graphite (Fig. 103) and very small spherulites distributed in an interdendritic pattern (Fig. 104). At high magnifications the spherulitic structure could just be discerned, and the spherulites were usually seen to be associated with very fine flakes (Fig. 105). The addition of calcium silicide to melts of this pure material

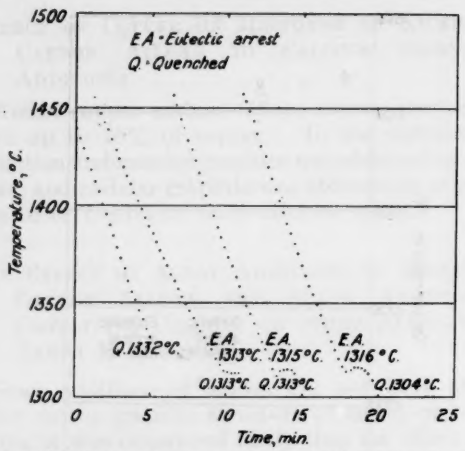

FIG. 136—Cooling curves of four cobalt–carbon ingots melted in graphite crucibles and quenched at different temperatures during solidification. (For structures *see* Figs. 106 to 110)

followed by air-cooling gave mixed structures of flake and undercooled graphite, as has been illustrated in Fig. 99.

Four cooling-curve experiments were carried out on 6-g. ingots, using the powdered cobalt melted in a graphite crucible. The first ingot was quenched in water before the eutectic arrest was reached, the second about half-way through the eutectic arrest, the third at 2° C. below the maximum temperature during the arrest, and the fourth at 12° C. below the maximum temperature of the arrest. The structures of these ingots are illustrated in Figs. 106, 107, 108, and 109, respectively, and the corresponding cooling curves are given in Fig. 136. These experiments do not reveal anything positive about graphite formation in cobalt–carbon alloys. The results are partly marred by the undoubted inoculating influence exerted by primary kish graphite. It is also impossible to say whether solidification was complete in the sample quenched at 2° C. below the maximum arrest temperature (Fig. 108). However, in Fig. 109, which shows the sample quenched at 12° C. below the maximum arrest temperature, an interesting feature is seen, a number of dark areas of very fine graphite being visible. It is doubtful whether this formed during solidification, and it does suggest a process similar to that in iron–carbon–silicon and nickel–carbon alloys. At high magnifications small amounts of cobalt carbide were found associated with this fine graphite (Fig. 110). In view of the fact that this alloy solidified with a largely coarse graphite structure it is conceivable that the alloy was not completely solid at the moment of quenching, although the cooling curve would seem to indicate otherwise.

Occurrence of Cobalt Carbide

The above experiments seemed to indicate that cobalt carbide is slightly more stable than nickel carbide. According to Ruff and Keilig[105] this carbide cannot be detected even in quenched samples. Meyer[106, 107] found that cobalt reacted with carbon monoxide to give, at 500–800° C., a carbide, Co_3C. The crystal structure of this carbide was found to be similar to that of Fe_3C. The unit-cell dimensions for these two compounds are compared below :

Axis	Co_3C (Meyer)	Fe_3C (Lipson and Petch)
a	4·52 Å.	4·5144 Å.
b	5·08 Å.	5·0787 Å.
c	6·73 Å.	6·7297 Å.

In view of the great similarity of the iron–carbon, the cobalt–carbon, and the nickel–carbon alloys which has been demonstrated and will be further buttressed, it seems strange that the structure of nickel carbide should be reported as close-packed hexagonal, $a = 2·646$ Å. and $c = 1·636$ Å., by Jacobson and Westgren[71] who prepared the compound from nickel powder and carbon monoxide.

It was found that by melting 6 g. of pure cobalt powder in a graphite crucible, heating to 1500° C., slowly cooling to 1400° C., and then quenching rapidly in water at about 15° C., cobalt carbide in the massive eutectic form and the Widmanstätten form could be obtained (Fig. 111). The samples always contained some graphite. The Widmanstätten carbide was undoubtedly deposited from the solid solution during the quench. It will be recalled that nickel carbide was obtained only in the Widmanstätten form by similar treatment. It would appear that cobalt carbide has a stability or persistence midway between that of iron carbide and nickel carbide. In other respects also, the properties of cobalt–carbon alloys lie midway between those of the iron–carbon (or rather iron–carbon–silicon) and the nickel–carbon alloys.

Metallography of Graphite in Cobalt–Carbon Alloys

Graphite flakes in the cobalt-carbon alloys exhibited the same optical properties as those in iron–carbon–silicon and nickel–carbon alloys. Between crossed nicols they lit up four times on revolving the stage through 360°, and under plane-polarized light they exhibited strong reflex pleochroism. The complex internal structures found in the graphite flakes of the other alloys were also present in the cobalt–carbon alloys. This is illustrated in Fig. 112.

Effect of Various Additions to Cobalt–Carbon Alloys

Additions of calcium silicide to cobalt–carbon alloys, even of the lowest sulphur content available, did not give nodular structures as with the nickel–carbon alloys but tended to give a coarsening of the structure. Additions of 0·5% of all the alloys listed in Table IV were made to hypereutectic cobalt–carbon melts, using graphite crucibles, and the results are summarized in Table V.

It is perhaps significant that every alloy containing magnesium gave some nodules and that the calcium–magnesium alloy was completely effective. Also noticeable was the fact that both

TABLE V—*Effect of Various Alloy Additions on the Structure of Hypereutectic Cobalt–Carbon Alloys*

Alloy Addition	Structure of Cobalt–Carbon Ingot
80% CaSi ...	Fine eutectic undercooled graphite. Flake kish floated to the top
Ca–Cu ...	A few nodules and remainder undercooled. Nodular kish at the top
Ca–Zn ...	Fine undercooled graphite with flake kish floated to top
Ca–Pb ...	Very coarse flake graphite
Sr–Mg ...	A few nodules and remainder undercooled. Nodular kish at top
Ba–Mg ...	Patches of nodules and patches of undercooled
Ba–Pb ...	Very coarse flake graphite
Ca–Mg ...	Completely nodular

additions containing lead gave coarse graphite. Fig. 113 shows the completely nodular structure of the cobalt–carbon alloy treated with calcium–magnesium; Fig. 114 shows the structure of the central regions of the ingot treated with calcium–copper; Fig. 115 shows the nodules at the top of this ingot (Figs. 114 and 115 are also typical of the ingot treated with strontium–magnesium); Fig. 116 shows the structure of the ingot treated with calcium–lead.

A cobalt–magnesium alloy was prepared containing approximately 30% of magnesium, and a hypereutectic melt was treated with 0·5% of this alloy. In the vicinity of the cobalt–magnesium compound in the resulting ingot, nodular graphite was produced. Again it appeared that alloyed magnesium will enable the graphite to form in the spherulitic pattern.

Temperature of Eutectic Arrest in Cobalt–Carbon Alloys

It will be recalled that Boecker[102] found the eutectic arrest at 1302° C. and Hashimoto[104] found it at 1315° C. In the work reported here the value 1321° C. was obtained for the untreated 40-g. ingot and, when using 6-g. ingots, 1313° C., 1315° C., and 1316° C. were the temperatures obtained. In all these cases the purer cobalt powder was used. The low value agrees with that of Hashimoto. The 40-g. ingot gave a coarse flake graphite structure, whereas the 6-g. ingots contained some finer graphite structures. If the finer graphite structures arise from undercooling, it is normal that the eutectic-arrest temperature should be lower than that of alloys with coarse flake graphite. It was suspected, however, that sulphur, apart from its effect on the graphite structure, would influence the temperature of the eutectic arrest.

The authors possessed no information about the cobalt–carbon–sulphur diagram, but the cobalt–sulphur binary diagram is well established.[108] It shows the occurrence of a cobalt/cobalt–sulphide eutectic solidifying at 879° C., with 26·5% of sulphur. It might be expected from this that sulphur additions to cobalt–carbon alloys would have a big effect in lowering the temperature of the eutectic arrest. In order to check this and to obtain a clearer knowledge of the true eutectic temperature in cobalt–carbon alloys, a number of cooling curves were taken on ingots weighing about 40 g., made up from cobalt powder and melted in graphite crucibles. Each ingot was slowly cooled from this temperature for the cooling curve. In each case a new thermocouple consisting of platinum and 10% rhodium–platinum wire of 0·25-mm. dia. was used. The e.m.f. of the couple was measured with a null-deflection potentiometer, and readings were taken to 0·005 millivolt. A summary of the more important of these experiments, with the structures obtained and the treatments used, is given in Table VI, and the corresponding cooling curves are given in Fig. 137.

From these results it is obvious that the sulphur content and the form of the graphite have a big influence on the temperature of the eutectic arrest in cobalt–carbon alloys. When coarse graphite is produced with a low sulphur content by means of the oxidized alkaline-earth alloys, the eutectic arrest occurred at 1319–1321° C., these being the lowest and highest values obtained by this treatment. This is in fair agreement with the previously obtained value of 1321° C. for an untreated ingot giving coarse graphite. When the coarse graphite was produced by the addition of sulphur, the eutectic arrest occurred at a considerably lower temperature, depending upon the sulphur content; *e.g.*, with 0·3% of sulphur it occurred at 1300° C. When an undercooled graphite structure was produced by melting with

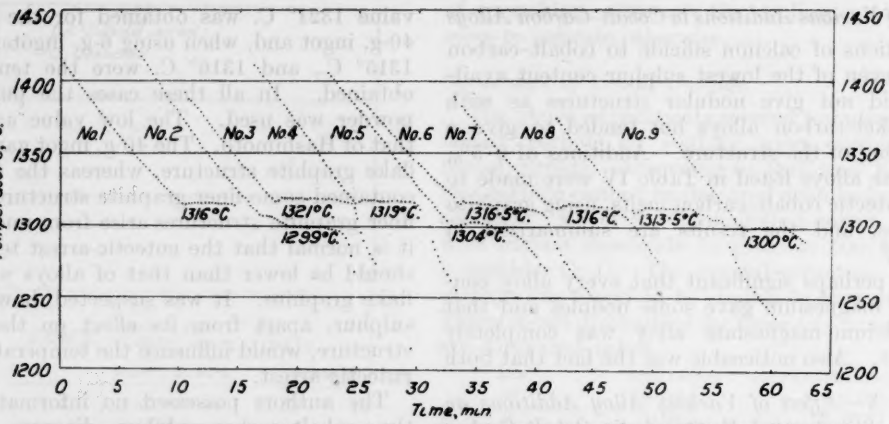

FIG. 137—Cooling curves of 40-g. cobalt–carbon ingots melted in graphite crucibles and treated in various ways to show the influence of sulphur content and graphite structure on the temperature of the eutectic arrest. (*See* Table VI and Figs. 117 to 119)

molybdenum sulphide, which added neither sulphur nor molybdenum to the final ingot, the arrest was at 1304° C. Fig. 117, taken from an ingot treated with strontium–magnesium alloy by ignition, shows a typical coarse graphite structure produced in this series, and Fig. 118 shows the undercooled graphite structure in the ingot treated with molybdenum sulphide.

It is felt that it can be concluded that in cobalt–carbon alloys of the same sulphur content, cooled at the same rate and similar in other respects, a coarse eutectic graphite structure may form at a temperature as much as 15° C. higher than that at which an undercooled graphite structure will solidify. This is in agreement with the idea of undercooling in these alloys and is exactly analogous to the behaviour of cast irons.

The structure produced by the addition of tellurium to a cobalt–carbon alloy is interesting. The authors have no knowledge of the cobalt–tellurium equilibrium diagram, but it may be assumed to be similar to the cobalt–sulphur diagram. If this is so, then tellurium additions might be expected to lower the temperature of

TABLE VI—*Temperature of Eutectic Arrest in Cobalt–Carbon Alloys*

Ingot No.	Treatment	Analysis	Structure	Temperature of Eutectic Arrest, ° C.
1	None	S, 0·012%	Mostly coarse flake with few patches of undercooled	1316
2	0·2% of Te added when molten	...	Abnormal undercooled graphite suggesting carbide breakdown. A few nodules	1299
3	0·5 g. of Sr–Mg alloy ignited on surface of molten alloy, and oxide then stirred in	S, 0·012%	Very coarse graphite	1321
4	0·5 g. of Ca–Mg alloy ignited on surface of molten alloy, and oxide then stirred in	S, 0·013%	Very coarse graphite	1319
5	0·2 g. of powdered molybdenum sulphide mixed with cold charge	S, 0·014%; Mo, nil	Kish + primary metal dendrites + undercooled graphite	1304
6	None	...	Coarse flakes + small patches of undercooled graphite	1316·5
7	0·2% of S, as roll sulphur, included in cold charge	S, 0·166%	Coarse graphite	1316
8	1% of Mn in charge. Treated with 0·3% of S when molten	S, 0·201%	Coarse graphite. Only manganese sulphide present	1313·5
9	0·5% of S, as roll sulphur, in cold charge	S, 0·320%	Coarse graphite. Cobalt sulphide	1300

the cobalt–carbon eutectic arrest. It has been shown that tellurium gives coarse graphite in nickel–carbon alloys, and that in cast irons it introduces a peculiar chilling effect arising from its carbide-stabilizing properties. The tellurium-treated cobalt–carbon alloys showed a structure (Fig. 119) which at low powers resembles undercooled graphite, but at high magnification was seen to be of the " mesh " variety, which has been described by one of the present authors elsewhere.[92] The " mesh " graphite structure is, in any given cluster, a continuous reticulated membrane of graphite. It has been suggested that it is produced in cast irons by the decomposition of a white iron after solidification, owing to a decrease in an external carbide-stabilizing influence (in cast irons this depended upon the

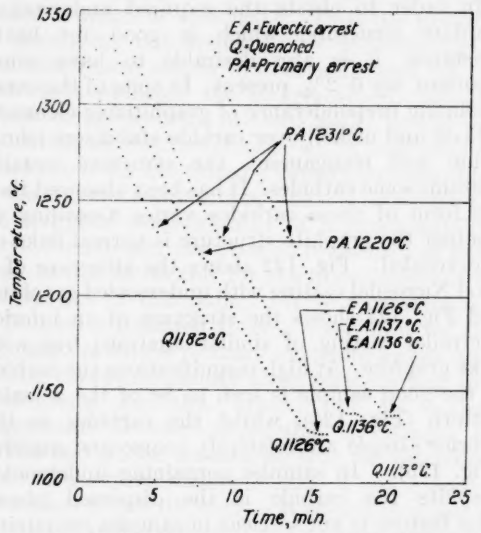

FIG. 138—Cooling curves of 7-g. cast-iron ingots containing 10% of silicon and quenched at various stages during solidification. (*See* Table VII and Figs. 120 to 122)

effect of temperature upon the reaction $MnS + Fe \rightleftharpoons FeS + Mn$). In the case of tellurium-treated cobalt–carbon alloys, it is possible that tellurium behaves as a carbide stabilizer, giving a eutectic carbide which decomposes after solidification. This is indicated to some extent by the simultaneous occurrence of a few spherulites in this structure. Furthermore, it has been found that cobalt–carbon alloys are susceptible to carbide stabilization by the addition of other carbide-stabilizing elements—considerable amounts of cobalt carbide have been obtained by the addition of sulphur or molybdenum to air-cooled cobalt–carbon ingots. If the addition of tellurium to ingot 2 (Table VI) did cause the solidification of a $Co–Co_3C$ eutectic, then this may have also contributed to its low eutectic arrest temperature.

It is relevant here to point out that undercooled graphite structures which have been produced, during this investigation in hypereutectic cobalt–carbon alloys were always associated with primary graphite crystals and primary metal dendrites. In this respect cobalt–carbon alloys closely resemble the cast irons. In nickel–carbon hypereutectic alloys large areas consisting entirely of undercooled graphite with no primary metal dendrites (Figs. 46 and 53) were easily obtained. No structures of this type have been found in cast irons or cobalt–carbon alloys, but, on the other hand, primary dendrites from both sides of the eutectic were occasionally obtained in nickel–carbon alloys (Fig. 45).

Formation of Undercooled Graphite in High-Silicon Cast Irons

It has been mentioned earlier that the presence of silicon or other graphitizing elements does not appear to influence the ease with which undercooled graphite can be produced in cast irons. Since this form of graphite is a product of carbide decomposition, one might expect factors influencing carbide stability to influence the form of its decomposition products. Hurst and Riley[109] have commented upon the occurrence of a carbide phase in high-silicon irons containing 10% and 15% of silicon. This phase appears to possess stability or persistence higher than one might expect from the mere extrapolation of the influence of silicon from lower values. These investigators, by planimetric measurements, indicated that the carbon content of this phase would be of the order of 2·0–3·5%.

To investigate the mode of formation of undercooled graphite in irons of high silicon content, a number of 0·875-in. dia. bars of the following composition were prepared :

T.C.	Si	Mn	S	P	Ti
1·10%	10·0%	0·72%	0·047%	0·028%	0·47%

The titanium was added to this iron in order to guarantee a fine undercooled graphite structure for relatively slow cooling rates. According to the work of Jass and Hanemann,[16] an iron–carbon–silicon alloy with this silicon content should have a eutectic carbon content of approximately 1·6%, the arrest beginning at about 1180° C.

Small ingots weighing about 7 g. each were machined from these bars and melted in the small silica crucibles which have been described and illustrated earlier. One sample was quenched whilst completely molten, one from a temperature

TABLE VII—*Quenching Experiments on High-Silicon Cast Iron*

Quenched	Temperature of Quenching, ° C.	Temperature of Primary Arrest, ° C.	Temperature of Eutectic Arrest, ° C.	Microstructure
From liquid	...	...	...	Carbide + δ ferrite
Between primary and eutectic arrest	1182	1231	...	,, ,, ,,
Before completion of eutectic solidification	1126	1220	1126	,, ,, ,,
On completion of euctetic solidification	1136	1231	1137	,, ,, ,,
Below eutectic temperature	1113	1231	1136	Undercooled graphite + δ ferrite

after the primary arrest but before eutectic solidification had begun, one during eutectic solidification, one immediately on completion of eutectic solidification, and one at 23° C. below the eutectic arrest. The cooling curves for the last four of these are given in Fig. 138. A summary of the microstructures and critical points in these four melts is given in Table VII.

The structure of the sample quenched immediately after solidification is shown in Fig. 120, and that of the sample quenched 23° C. below the eutectic arrest is shown in Fig. 121. It may be argued that, from the cooling curve, it was impossible to say the alloy quenched at 1° C. below the arrest was completely solid, but without doubt the eutectic crystallization had occurred without the formation of graphite. Since Fig. 121 shows normal undercooled graphite, it follows that these alloys solidify white and subsequently graphitize to give undercooled graphite. The dark spots seen in the micrograph are titanium carbide, standing in relief.

It should be noted that in no case was the eutectic arrest a horizontal—this is to be expected in alloys of this composition—and the arrest temperatures quoted are approximately the mid-temperatures of the arrest. On these cooling curves it was impossible to assign definite values for the beginning and ending of eutectic crystallization, but the values are considerably lower than those to be expected from the work of Jass and Hanemann. This may be owing to the presence of impurities (notably manganese and titanium) or to the German workers' values being determined on alloys of exactly eutectic composition, or to the fact that they did not obtain undercooling.

THE FORM OF THE CARBIDE IN AUSTENITIC CAST IRONS

The proprietary cast iron Nicrosilal is a heat-resisting iron which, when correctly manufactured, should have a structure of undercooled graphite with carbides in a matrix of austenite. The composition of this alloy should fall within the following range:

T.C. %	Si, %	Mn, %	P, %	Ni, %	Cr, %
1·6–2·6	4–5	0·5–1	<0·15	18–20	2–4

In order to obtain the required undercooled graphite structure, which is good for heat-resistance, it is also desirable to have some titanium, say 0·2%, present. In spite of the overwhelming preponderance of graphitizing elements (silicon and nickel) over carbide stabilizers (chromium and manganese), the structure usually contains some carbides. It has been observed that the form of these carbides varies according to whether the graphite structure is normal flake or undercooled. Fig. 122 shows the structure of a good Nicrosilal casting with undercooled graphite, and Fig. 123 shows the structure of an inferior Nicrosilal casting of similar analysis, but with flake graphite. At high magnifications the carbide in the good sample is seen to be of the acicular pattern (Fig. 124), whilst the carbides in the inferior sample are relatively coarse and massive (Fig. 125). In samples containing undercooled graphite the carbide is the dispersed phase. This feature is not so clear in samples containing flake graphite, but certainly in many spots the carbide is the continuous phase of the eutectic.

Obviously in this case we have eutectic carbide occurring in two differing forms, depending upon whether the alloy has undercooled or not. In this context it is interesting to recall that Norbury[30] postulated the existence of a normal and undercooled white iron.

UNDERCOOLING—A RECAPITULATION OF IDEAS IN RELATION TO THE PRESENT PROBLEM

It has now been clearly demonstrated for a wide range of iron–carbon–silicon alloys, nickel–carbon alloys, and cobalt–carbon alloys, that undercooling can occur. The normal flake graphite forms by the solidification of a eutectic or binary complex of solid solution and graphite, whereas the undercooled structure in all three alloys probably results from the solidification of a

binary complex of solid solution and carbide, which decomposes to give a fine form of graphite at some temperature after complete solidification.

The concept of a supersolubility curve for eutectiferous and other binary mixtures has been developed by Miers [110] and his co-workers for certain organic compounds. Gayler [111] has applied this concept to the aluminium–silicon alloys and has determined the position of the supersolubility curve for this system, using rapidly cooled and sodium-modified alloys. This investigator not only located the hypertectic point, but was able to plot the supersolubility curve in the hypo-eutectic and hypereutectic regions. In many respects the aluminium–silicon alloys possess a remarkable resemblance to the iron–carbon–silicon alloys in that solidification of the eutectic at normal temperatures gives silicon in a lamellar form, whereas, by undercooling, a finely dispersed form exactly similar in appearance to undercooled graphite is obtained. Thus the problem of correlating our knowledge of undercooling in the aluminium–silicon system with that of the iron–carbon–silicon, the nickel–carbon, and the cobalt–carbon systems presents itself. Gayler was able to determine the extent to which undercooling could occur for the primary arrests on both sides of the eutectic composition. This is necessary for the complete establishment of a supersolubility curve, but, so far, the present investigators have been unable to do this for the alloys relevant to the subject of this paper. Wide variations are undoubtedly obtained from a given iron for the temperature of the primary austenite arrest, and cooling curves frequently show undercooling and recalescence for this arrest, but its actual temperature is very susceptible to changes in carbon and silicon content, so that the establishment of a supersolubility curve—if any exists—is a matter of real difficulty. It should be possible to prevent undercooling at the primary arrest by the addition of austenite just before solidification begins. In this connection it is of significance to note that, as far as the solidification of the eutectic is concerned, inoculation apparently occurs only with graphite additions or in the presence of kish graphite, and not with austenite.

In her discussion on the solidification of undercooled hypo-eutectic aluminium–silicon alloys, Dr. Gayler has suggested the following sequence: Aluminium separates at a point on the supersolubility curve, and continues to be deposited at constant or rising temperature; the silicon content of the remaining liquid is thereby increased until the normal solubility curve is reached, down which aluminium is deposited until the silicon-supersolubility curve is reached, and silicon is then deposited until the hypertectic point is reached. This explanation, if modified for cast irons, would account for the arrest and recalescence noted at the beginning of austenite separation in the cooling curves for some cast irons.

Thus far it seems that the system aluminium–silicon and those under consideration in this paper are analogous. It should, however, be noted that whereas undercooled graphite solidifies as a carbide structure, we have so far not found analogous behaviour in aluminium–silicon alloys. In all the considerations in the literature of undercooling according to the betol–salol example in non-metallic and metallic eutectics, the present investigators have never found any reference to the possibility of unstable compounds solidifying at the hypertectic, with subsequent decomposition after solidification. Is the behaviour of iron–carbon–silicon, nickel–carbon, and cobalt–carbon alloys unique in this respect, or can the concept of the formation of metastable compounds be extended to other systems? This is an absorbing question the solution of which is not an easy matter. The present investigators have had no success in proving the existence of a metastable phase in aluminium–silicon alloys, but it is suggested that the possibility should be rigorously examined.

One further type of undercooling in metallic systems[112] should be mentioned here. When a system is such that an alloy will begin to deposit compound *A* from the liquid until, at a peritectic transformation, *A* reacts with the liquid to give compound *B* and eventually, when this transformation is complete, a eutectic of compound *B* and another phase, *C*, is obtained, the following may occur. In the absence of nuclei of *B* the alloy will undercool beyond the peritectic transformation, with the deposition of *A* until a metastable eutectic of *A* and *C* is obtained, which may subsequently decompose after solidification to give *B*. There are various modifications of this sequence for different types of systems, but in every case a metastable eutectic is formed which decomposes after solidification. Examples of this are the zinc–antimony system,[113] the cadmium–arsenic system,[114] and the iron–phosphorus system.[115] In the zinc–antimony system a maximum on the liquidus curve corresponds to Zn_3Sb_2. The compound ZnSb is formed by a peritectic reaction between Zn_3Sb_2 and antimony, but it does not make its appearance in the absence of inoculation. A metastable eutectic of Zn_3Sb_2 and antimony may decompose as follows:

$$Zn_3Sb_2 + Sb \longrightarrow 3ZnSb.$$

In the sense of the formation of a metastable eutectic which decomposes after solidification, this type of undercooling is analogous to that in the iron–carbon–silicon alloys, but in the light of our knowledge of the liquidus surface in these ternary alloys the analogy cannot be exploited to explain undercooling in cast irons.

Summary and Conclusions

A fairly comprehensive account has now been given of a large number of problems which either need explanation or correlation in connection with the formation of graphite in grey cast iron. Ultimately it should be possible to do this in relation to a single unified theory covering all the various phenomena. In spite of the large numbers of debatable points dealt with, it is felt that many more exist and need to be incorporated into the whole as soon as possible. Only by considering the graphite-formation problem in all its many aspects are we likely to achieve a satisfactory general explanation and hence complete metallurgical control of the material.

In this paper a deliberate attempt has been made to present the facts as given in the literature, interwoven with observations made in the course of the work of the present investigators. The actual experimental work reported here covers only a fraction of the total field to be investigated. However, it is felt that certain facts have been uncovered which enable a bolder approach to be made to the problem. No work has been reported by the present investigators on the effect of gas content or slag–metal equilibria. This section will be tackled next, and it is hoped that it will form the subject of a further report in the future.

Whilst observations have been made and experimental evidence produced on a number of diverse issues, the main weight of the paper has been directed along the following lines:

(*a*) To study and develop the analogies existing between the process of graphite formation in cast irons on the one hand, and in nickel–carbon and cobalt–carbon alloys on the other.

(*b*) To confirm the findings of Eash that undercooling in cast irons occurs with the solidification of a white-iron structure which subsequently decomposes after solidification to give undercooled graphite, and to extend this idea to high-silicon irons and to nickel–carbon and cobalt–carbon alloys.

(*c*) To examine the possibility of nodular graphite occurring in as-cast cast irons, *i.e.*, without any heat-treatment.

These three lines of investigation are closely related at several points and cannot be separated for consideration, but the following main conclusions can be drawn and the following suggestions made, bearing in mind all three:

(1) The analogies existing between the formation of graphite in cast irons, the nickel–carbon alloys, and the cobalt–carbon alloys, can be developed to a surprising degree of detail.

(2) A study of graphite formation in nickel–carbon and cobalt–carbon alloys can yield useful information about the same process in the iron–carbon–silicon alloys and in cast irons.

(3) Whilst there are bound to be some differences of detail, it is reasonable to suppose that any general theory of graphite formation in cast irons may be equally applicable to nickel–carbon or cobalt–carbon alloys. Application in this manner may in fact prove a useful test for the validity of a theory. There is an obvious limit to this process, and it must be used with caution.

(4) Nickel–carbon and cobalt–carbon alloys yield graphite structures which may be of the normal flake variety or of the undercooled variety.

(5) In both nickel–carbon and cobalt–carbon alloys, the eutectic giving coarse flake graphite solidifies at a higher temperature than that giving undercooled graphite, provided that the sulphur content is the same in each case.

(6) The presence of sulphur (and/or tellurium in the case of nickel–carbon alloys) in nickel–carbon and cobalt–carbon alloys tends to prevent the formation of undercooled structures giving coarse flake graphite. In this respect these alloys show an interesting resemblance to iron–carbon–silicon alloys. This function of sulphur is not understood, but it behaves in this manner in nickel–carbon alloys, whether present as nickel sulphide or as manganese sulphide.

(7) Nickel–carbon and cobalt–carbon alloys are susceptible to inoculation by graphite.

(8) Nickel–carbon and cobalt–carbon alloys are either directly or indirectly susceptible to inoculation by silica.

(9) The addition of calcium (and also probably barium and strontium) or magnesium can cause the formation of spherulitic nodular structures in nickel–carbon and cobalt–carbon alloys, provided that the sulphur content of the alloys is sufficiently low. The calcium and magnesium may be added to the alloys in a wide variety of forms in combination with other elements if necessary. These structures are produced with greater ease in nickel–carbon alloys than in cobalt–carbon alloys.

(10) Rapid cooling of nickel–carbon alloys can also give nodular graphite structures.

(11) Eutectic and hypereutectic graphite can be obtained in the spherulitic form in nickel and cobalt–carbon alloys.

(12) The existence of nickel carbide (probably Ni_3C) has been shown in nickel–carbon alloys, and cobalt carbide (probably Co_3C) in cobalt–carbon alloys.

(13) From this work it would appear that Fe_3C is more persistent (or "stable") than either Co_3C or Ni_3C, and that Ni_3C is less persistent than Co_3C. In many respects cobalt–carbon alloys show properties midway between those of iron–carbon(–silicon) and nickel–carbon alloys. For instance, nodules are produced with ease in nickel–carbon alloys, with a little difficulty in cobalt–carbon alloys, and only with great difficulty in iron–carbon–silicon alloys. Cobalt–carbon alloys are probably susceptible to carbide stabilization by the addition of elements such as tellurium; this is also the case in iron–carbon–silicon alloys, but not in nickel–carbon alloys. That there should be a gradation in properties going through the series Fe, Co, Ni, is understandable. Whether all the observations which have been made merely relate to differences of degree and are controlled by the similar electronic structure of the elements iron, cobalt, and nickel, it is as yet impossible to say. It is conceivable that a more exhaustive study will reveal qualitative differences in behaviour between these three groups of alloys. If this is the case, the differing crystal structure of nickel carbide compared with the cobalt and iron carbides may be a determining feature.

(14) The effect of calcium silicide additions to nickel–iron–carbon alloys in giving spherulitic structures has been shown to persist for iron contents of up to about 50%, beyond which the normal graphite-coarsening influence is obtained.

(15) The work of Eash in showing that undercooled graphite structures in cast irons result from the decomposition of a white-iron structure has been confirmed and shown to be applicable even in the presence of very high silicon contents. That normal flake graphite forms directly from the liquid has been confirmed by the present investigators on many occasions.

(16) The form of the white-iron structure which is responsible for the formation of undercooled graphite is acicular in appearance and differs from normal white-iron eutectics in this respect.

(17) Undercooled graphite in nickel–carbon alloys most probably forms in a similar manner, *viz.*, a eutectic of a solid solution of carbon in nickel and nickel carbide solidifies, and the nickel carbide decomposes after solidification, to give undercooled graphite.

(18) Similarly, in cobalt–carbon alloys undercooling may occur with the solidification of a eutectic of solid solution and cobalt carbide which subsequently decomposes after solidification.

(19) The freezing point of the eutectic in calcium silicide treated nickel–carbon alloys giving spherulitic graphite is substantially the same as that in alloys giving undercooled graphite.

(20) It is suggested that the action of the elements calcium and magnesium in giving nodular structures in nickel–carbon alloys is due to these elements causing some stabilization of the nickel carbide, so that it decomposes at a lower temperature than that required to give undercooled graphite. A similar explanation would explain the similar effects in cobalt–carbon alloys. This explanation agrees with the finding that rapid cooling also gives nodules in nickel–carbon alloys; under these conditions also the nickel carbide would not have time to decompose until lower temperatures had been reached.

(21) A search has shown that spherulitic nodular structures can also occur in cast iron. Broadly speaking, these occur in two different ways, *viz.*:

(*a*) In hypereutectic alloys which have undercooled. This undercooling may be induced either by rapid cooling or by the removal of "nuclei" from the melt, *e.g.*, by vacuum melting.

(*b*) In hypo-eutectic alloys associated with inverse chill. It has been shown that inverse chill may be classified into two types and that one type possesses the acicular white-iron eutectic structure which normally decomposes after solidification to give undercooled graphite. The nodules occur associated with this form, and it has been suggested that they are produced by the decomposition of the acicularly formed cementite at some temperature below that required to give undercooled graphite. If this is correct then the process may be similar to that occurring in nickel–carbon alloys.

(22) No evidence has been obtained in any of the alloys studied that spherulitic graphite structures form directly from liquid. It appears that a solid carbide must decompose to yield a

spherulitic structure, and this confirms experience in the malleablizing process.

(23) Deliberately to produce spherulitic structures in cast irons it is necessary to cause an iron to solidify white and then for the cementite to decompose after solidification. It has been shown that decomposition of carbide after solidification can be produced by suitable use of the reaction $FeS + Mn \rightleftharpoons MnS + Fe$. Except in special cases, this type of carbide decomposition does not yield spherulites, and then only part of the total graphite content is in this form. Without any provision of nuclei the carbide will decompose to give a "mesh-like" graphite pattern.

(24) The production of completely spherulitic structures in cast iron requires an iron which will undercool to give the acicular white-iron structure on solidification and the presence of an alloying element, which will cause it to persist until a temperature probably within the range 800–1050° C., beyond which the carbide would be permitted to decompose.

(25) For an iron of any given composition with respect to the elements carbon, silicon, manganese, sulphur, and phosphorus, it is possible that two forms of white iron exist, *viz.*, one (the normal white-iron structure) which, once it has formed, has an increasing tendency to persist (stability) with falling temperature, and one which tends to decompose after solidification to give undercooled graphite and which has an acicular-distributed form of eutectic cementite. It is possible that the second form corresponds to the maximum degree of undercooling obtainable in an iron.

It is felt that the conclusions and suggestions listed above are all to some extent interrelated and reveal an aspect of the metallurgy of cast irons not hitherto dealt with. In addition to this, many observations have been introduced into this paper which are normally dealt with individually but which must be generally borne in mind in order to explain the process of graphite formation in cast irons. A few of these may be listed as follows :

(i) The relation between the graphite-refining influences of titanium and zirconium and the form of the minor phases necessary to give the effect (TiC or ZrC is necessary).

(ii) The influence of tellurium.

(iii) The graphite-refining effect of bismuth.

(iv) Effect of sulphur in promoting flake graphite formation.

(v) Inoculation with ferrosilicon.

(vi) Effect of hydrogen in giving flake graphite.

(vii) Influence of superheating.

(viii) Effect of slag composition.

(ix) Formation of inverse chill.

If these features alone could be introduced into one unified theory, a considerable advance would be made. However, many other facts also require explanation.

In the course of this work and whilst writing this paper, liberal use has been made of the terms "nuclei," "undercooling," and "spherulitic structures." The present investigators are impressed that these terms touch on spheres of relatively abysmal ignorance. The term "nucleus" is widely used, but its precise connotation is rarely defined and it may refer to small solid particles, at one extreme, or merely to a physical or chemical condition favouring a particular type of crystallization at a given point, at the other extreme. Between these two concepts we have the idea of "residual lattices" in liquids where some of the atoms may be linked together in a similar manner to that of the solid. These aggregates could provide centres for the initiation of crystallization on subsequent cooling, or alternatively could be completely destroyed by superheating. Some visible solid particles can apparently function to nucleate a melt in the sense of preventing undercooling. It would be interesting to know to what extent the crystal structure of such particles influences their ability to initiate crystallization and to prevent undercooling. At times, however, "nucleation" is spoken of when no visible solid particles are in evidence. It is a useful concept to some extent, but knowledge of the nature of the "nuclei" must precede a fuller understanding of the process of solidification in cast irons, and also in other metallic and non-metallic substances and mixtures.

As far as undercooling during solidification is concerned, published work consists almost entirely of the classical researches of Tammann and his co-workers in Germany and Miers and his co-workers in this country. For the most part these workers have used non-metallic substances for study, and undercooling in metallic systems has only been approached at all systematically in the case of aluminium–silicon alloys and cast irons. Further work on the fundamentals of undercooling in non-metallic and metallic mixtures is urgently required. The process is of immense practical significance in metallurgy, because the useful properties of an alloy may be considerably modified by the promotion or prevention of undercooling during solidification.

The existence of graphite in a spherulitic pattern in iron–carbon–silicon, nickel–carbon, and cobalt–carbon alloys is a unique feature, not

only for graphite but also for metallic systems, as far as the present authors are aware. What determines whether crystallization or the formation of a new phase shall give a spherulitic pattern ? Can spherulites form directly from a liquid or is a constricting condition necessary, such as that existing in a transformation in the solid or the solidification of liquid crystals ? Is it necessary that a particular type of nucleus be present to initiate spherulitic crystallization and, if so, does the crystal structure of this nucleus play a part ? Spherulitic graphite formation during the malleablizing process has been shown by one of the authors to occur only in the presence of iron sulphide, and frequently iron sulphide occurs at the centres of graphite spherulites. Manganese sulphide does not have this influence. Is this owing to a difference in crystal structure ? From the authors' appreciation of the literature on this subject it appears that very little is known, but the subject is of vast importance to the metallurgy of cast iron because it involves the difference between a brittle non-ductile material and one with shock resistance and ductility. Again, one might ask what might be the properties of other alloys if other phases in other systems could be made to crystallize spherulitically. For instance, it is conceivable that aluminium–silicon alloys could be produced with vastly different properties if the eutectic silicon could be obtained in this form. It may be that only substances with a certain type of crystal structure can be made to crystallize spherulitically, and this would limit the possibilities. The field here is rich for further fundamental work.

Finally, the present investigators would like to emphasize the value in physical-metallurgical research of studying a particular type of problem in a number of alloys of different components, rather than in one alloy only. It is felt that this opinion is supported by some of the results presented in this paper. The analogous systems must be chosen judiciously. One element at least must be common to the systems, or one element in each system should possess a common feature, such as electron arrangement, and, better, still, in essentially binary alloys one element should be common in each system and the other elements should possess common features. This has been forecast and demonstrated by the theoretical physical metallurgist for many years.

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Historical Note No. 1

Housing Conditions of Ironworkers in the Sixteenth Century*

By Dr, H. R. Schubert†

The sixteenth century was a most important period in the history of iron and steel making in Britain, as it was marked by a great expansion which affected almost all branches of the industry. There is abundant information regarding ironworks, output, finance, and the social and economic position of the ironmasters of the time, but much less is known about the living conditions of the workers. A fair amount of documentary evidence offering insight into the wages earned by the labourers in the iron industry exists, but hardly anything is mentioned regarding housing.

When an ironworks was erected at or near an abbey or priory sold into private hands after the suppression of the monasteries, there was no difficulty in finding suitable accommodation for the housing of staff and labourers in the various buildings of such places. When Sir Henry Sidney, the father of the famous poet, had a steelworks erected at Robertsbridge (Sussex) in 1565, the steelmakers were lodged in houses belonging to the former abbey, and this arrangement was still in force in 1609.[1] On the other hand, we sometimes learn of ironworks which had no accommodation for the workers at all. This was the case at the forge near Goodrich (Herefordshire) in 1544, where the ironmaster had to board his four ironworkers in the neighbouring village of Whitchurch, but when he made no

* Received 18th February, 1947.
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[1] E. Straker, "Wealden Iron" (second edition), p. 314. London 1931: G. Bell and Sons, Ltd.

high phosphorus contents, *i.e.*, 0·010% and 0·030% of phosphorus, with and without the addition of 1% of titanium.

In general, low results for phosphorus were recorded, although varying degrees of interference due to this element were obtained by the individual

Effects of Deoxidation Practice. (Steel, 1945, vol. 117, Dec. 3, pp. 114–117, 154). Prolonged service of steel at elevated temperatures depends on the microstructure, and this itself changes throughout the service life. These microstructural changes occurring over longer periods at high temperature have been studied for three carbon and six molybdenum steels. A relationship between spheroidization and graphitization in the presence of the aluminium added for deoxidizing is probable. Tables show how spheroidization and graphitization progress in steel at various temperatures with increasing aluminium additions and with different deoxidation practices.

The Reduction of Silica by Liquid Steel. W. Geller. (Revue de Métallurgie, Mémoires, 1943, vol. 40, June, pp. 183–191; July, pp. 219–224). This is a French translation of a paper which appeared in Archiv für das Eisenhüttenwesen, 1942, vol. 15, May, pp. 479–490 (*see* Journ. I. and S. I., 1945, No. II., p. 184 A).

Note on Oxide Systems Pertaining to Steel-Making Furnace Slags. A. H. Jay and K. W. Andrews. (Journal of the Iron and Steel Institute, 1945, No. II., pp. 15 P–18 P). An X-ray survey has been made of the binary systems FeO–MnO, FeO–MgO, CaO–MnO, and MgO–MnO. All systems show complete solubility at a temperature of 1150° C. The variation of lattice spacings with composition in each system is recorded.

Electric Steel-Making in the Arc Furnace. C. Wissmann. (Electrochemical Society, Oct., 1945, Preprint No. 12). Descriptions are given of the acid and basic processes of making steel in the electric-arc furnace.

Some Thoughts on Experimental Electric Arc Furnace Smelting. H. S. Newhall. (Electrochemical Society, Oct., 1945, Preprint No. 13). Electric-arc furnace design, choice of phase, electrodes, refractories and the operation of experimental steelmaking furnaces are discussed.

Electric Induction Steel "E.I.S." F. T. Chesnut. (Electrochemical Society, Oct., 1945, Preprint No. 17). A review is presented of the present status of the application of the high-frequency induction furnace to steel-melting.

A Thermal Engineer's View of an Arc Furnace. V. Paschkis. (Electrochemical Society, Oct., 1945, Preprint No. 20). Means of increasing the melting efficiency of the electric-arc furnace are considered from the heat-transfer point of view.

Steel-Making; A Modern Installation of Arc Furnaces. (Electrical Review, 1945, vol. 137, Dec. 7, pp. 809–814). A steelmaking plant comprising three Siemens 15-ton and one 5-ton electrode furnaces and accessory equipment at the works of the Brymbo Steel Company, Ltd., North Wales, is described.

Safety Practices. R. L. Beeswy. (Iron and Steel Engineer, 1945, vol. 22, Oct., pp. 77–80). The safety precautions relating to electrical machinery and circuits, and especially to cranes, at the works of the Inland Steel Company are described.

FORGING, STAMPING, AND DRAWING

(Continued from pp. 5 A–6 A)

World's Largest Hydraulic Presses. (Steel, 1945, vol. 117, Dec. 3, pp. 130–186). A description is given of two hydraulic presses built to forge armour plate, hollow forgings, and shafts for the American Navy. The maximum pressing force is 14,000 tons eccentric load capacity 6 ft. off centre; the working stroke is 10 ft. and the overall height 74 ft. Controls permit the press head to be moved to within $\frac{1}{16}$ in. of any desired position.

Magnetic Core Laminations Produced at Minimum Cost. G. M. Shingledecker. (Machinist, 1945, vol. 89, Dec. 29, pp. 2125–2127). The fabrication of magnetic core laminations is essential for the manufacture of electric motors. All laminations must conform to rigid standards of dimensional accuracy and magnetic performance so that the selection of a good tool steel with correct stamping qualities, the controlled annealing of the laminations to obtain the desired electrical characteristics, and the use of progressive dies of sectional type are necessary for ideal lamination production.

Forging Die Design. J. Mueller. (Steel Processing, 1945, vol. 31, Oct., pp. 633–635). Recommendations are made on the use of inserts in forging dies for mass-production work.

Factors Affecting Die Life. A. S. Jameson. (Steel, 1945, vol. 117, Nov. 26, pp. 101–106). A statistical analysis is made of factors affecting the life of the dies used for cold-heading in the manufacture of bolts with square necks.

Estimating Die Costs. R. Howell. (Iron Age, 1945, vol. 156, Nov. 29, pp. 44–48). The differences which frequently prevail in estimating the costs of forging dies are due in only a small degree to variations in labour and material costs, but may well be the result of faulty or inefficient recording methods. In this article the author presents the costing methods which are employed by a company making large numbers of drop-forgings.

Recent Engineering Developments Contribute to Greater Stamping Production. (Steel Processing, 1945, vol. 31, Oct., pp. 636–638). Examples are given of the manner in which machine-gun mountings and other small ordnance parts were produced in large numbers by improvements to stamping processes.

High Forging Temperatures Revealed by Facets in Fracture Tests. J. R. Strohm and W. E. Jominy. (American Society for Metals, Oct., 1945, Preprint No. 25). The occurrence of large grains or light-coloured facets in the fractures of forged low-alloy steel bars is discussed. They indicate that the forging temperature was too high. The critical temperature below which these facets did not appear was determined for a number of steels.

Forming Tubular Products and Closing Tube Ends by Flame Spinning. (Steel, 1945, vol. 117, Nov. 26,

pp. 92–93). A process of forming the ends of tubes and hollow bodies is described. The tube end is rotated under an oxy-acetylene multiple-tip burner and the closing is done with a hard-faced forming tool or shoe.

Manufacture of Fine Steel Wire and Some of Its War Applications. J. R. Thompson. (American Iron and Steel Institute : Blast Furnace and Steel Plant, 1945, vol. 33, Oct., pp. 1248–1252, 1276, 1277).

Coatings and Lubricants. (Wire Industry, 1945, vol. 12, Nov., p. 575). The characteristics and limitations of lime as a wire-drawing lubricant are discussed.

ROLLING-MILL PRACTICE

The Rolling of Metals: Theory and Experiment, L. R. Underwood. (Sheet Metal Industries, 1945, vol. 22, Aug., pp. 1360–1366 ; Sept., pp. 1535–1545 ; Oct., pp. 1719–1724, 1736 ; Nov., pp. 1905–1913 ; Dec., pp. 2089–2095). The discussion of the interaction of roll diameter, dimensions and nature of material, etc., in the rolling of metals is continued in Parts V., VI., and VII., which deal respectively with the factors affecting the yield stress in cold- and hot-rolling, external friction between the rolls and the material, and factors influencing rolling load and specific roll pressure. (*See* Journ. I. and S. I., 1945, No. II., p. 110 A).

Rail and Structural Mill for Volta Redonda Plant Designed for Varied Output. C. Longenecker. (Blast Furnace and Steel Plant, 1945, vol. 33, Sept., pp. 1108–1112). A description is given of the rolling-mill plant which will shortly be put in commission at the Volta Redonda steelworks in Brazil.

Bethlehem Centralizes Bar Mills. (Iron Age, 1945, vol. 156, Nov. 15, p. 77). A continuous high-speed 10-in. merchant bar mill, with a monthly capacity of 22,000 tons in sizes of $\frac{5}{16}$ to $1\frac{1}{4}$ in. rounds at one of the Bethlehem Steel Company's works is described.

Rolling of Large Diameter Tubes. (Machinery, 1945, vol. 67, Nov. 22, pp. 568–570). A mill is described for rolling pierced billets to construct hollow bodies up to 40 ft. long and from 4 to 6 ft. in dia.

Electric Equipments for a 16-Stand Tube Mill. A. L. Thurman. (Blast Furnace and Steel Plant, 1945, vol. 33, Oct., pp. 1235–1241, 1247). Descriptions of modern tube-reducing mills are given. An example is a 16-stand mill with the stands only $12\frac{1}{2}$ in. apart. This is capable of reducing $4\frac{1}{2}$-in. dia. tubing down to $\frac{3}{4}$ in. in dia., the exit speed being 300 to 500 ft./min.

The Manufacture of Butt-Welded Pipe. N. W. Richardson. (Iron and Steel Engineer, 1945, vol. 22, Nov., pp. 64–68). The Fretz-Moon process of making pipes by continuous butt-welding and the requisite plant are described and illustrated. Important advantages are reduced costs of operation and maintenance, and higher yield and quality. Mills of this type are in operation in the United States, England, Australia, Russia, Germany, and Canada.

Improving Production in the Finishing Shop of a Plate Mill by Work- and Time-Studies on the Straightening Rolls. Part II. E. Kratschmar. (Archiv für das Eisenhüttenwesen, 1942, vol. 16, July, pp. 39–44). Continuation of a previous article. (*See* Journ. I. and S. I., 1943, No. II., p. 82 A).

Work-, Time-, and Cost-Studies, Explained by the Example of the Cutting of Plates with Hydraulic Shears. Part III. E. Kratschmar. (Archiv für das Eisenhüttenwesen, 1942, vol. 16, Sept., pp. 117–127). Conclusion of a series of articles (*see* preceding abstract).

Rolling Mill Guides. W. F. Hoffman. (Iron and Steel Engineer, 1945, vol. 22, Nov., pp. 69–79). Problems in the design of entry and exit guides for rolling mills are discussed and many designs of guides are described and illustrated.

Cooling Beds and Transfers. C. L. Raisig. (Iron and Steel Engineer, 1945, vol. 22, Nov., pp. 37–47). Several modern cooling beds for rolled bars are described in detail and the requirements for an efficient bed are discussed. They should be flexible enough to meet all operating demands. Useful data are given correlating the dimensions of various sections and the cooling times required.

Electronic Controls. F. Roberson. (Iron and Steel Engineer, 1945, vol. 22, Oct., pp. 59–61). Some applications of electronic devices for controlling machinery in the iron and steel industry are described and illustrated.

Rectifiers in the Steel Industry. F. Mohler. (Iron and Steel Engineer, 1945, vol. 22, Oct., pp. 66–74). The theory of current rectifiers is explained with descriptions of phanotron or thyratron, ignitron, and metal plate rectifiers and their applications in the steel industry.

Continuous Gaging with X-Ray Micrometer. R. C. Woods and F. Fua. (Iron Age, 1945, vol. 156, Nov. 29, pp. 50–51). The thickness of sheet in motion can be gauged with high sensitivity by a device which combines X-rays and electronic instruments. The method described can be applied to measure the wall thickness of long metal tubes at any point, and in the case of coated metals the thickness of the coating and that of the base metal can be measured independently.

PYROMETRY

Some Practical Aspects of Pyrometry. R. G. Ackland. (Australian Institute of Metals : Australasian Engineer, Science Sheet, 1945, Sept. 7, pp. 9–16). The theory and practice of pyrometry is surveyed with descriptions of thermocouples, millivoltmeters, potentiometric instruments, optical pyrometers and methods of checking instruments.

Pyrometry. H. J. Smith. (Metal Industry, 1945, vol. 67, Dec. 14, pp. 395–397 ; Dec. 21, pp. 415–417). Developments in the design of new pyrometric

equipment and its application to industrial plant during the last six years are reviewed.

The Design and Performance of Some Commercial Optical Pyrometers of the Disappearing-Filament Type. C. R. Barber. (Journal of the Iron and Steel Institute, 1945, No. II., pp. 171 P–188 P). A critical examination has been made of five models of commercial portable optical pyrometers of the disappearing-filament type, with reference to the design of the lamp, the optical system, the monochromatic and neutral filters, and the measuring system.

It has been shown that in many cases the lamp filaments are too short to be free from end-effect, and the advantages of the use of a flat-filament in this and other respects are discussed. The size and position of the diaphragms of the optical system were in some examples not in accordance with those necessary for obtaining perfect disappearance of the filament. The transmission curves of the monochromatic and neutral filters are given, and the requirements of the combination of the two filters are examined. A current or voltage calibration of the lamp may be employed and the advantage of the latter for short filaments is pointed out. The potentiometer method used in one pyrometer has considerable merit.

The Measurement of Heat Flow from Furnace Walls. C. B. Bradley. (American Society of Mechanical Engineers: Industrial Heating, 1945, vol. 12, Aug., pp. 1318–1320). Methods of measuring the flow of heat through a furnace wall are discussed and a device consisting of two thermocouples mounted in bakelite for fixing to slabs of insulating material is described; this is used to measure the insulating efficiency of the slab. Practical difficulties encountered with this method are pointed out.

HEAT-TREATMENT

(Continued from pp. 6 A–8 A)

The Application of Ms Points to Case Depth Measurement. E. S. Rowland and S. R. Lyle. (American Society for Metals, Oct., 1945, Preprint No. 24). A method of measuring case depth has been described previously by the authors (*see* Journ. I. and S. I., 1945, No. II., p. 111 A). In the present paper the results obtained when applying the method to a number of low-alloy chromium-nickel steels with carbon ranging from 0·40% to 1% are presented and discussed.

Time-Temperature Transformation Curves for Use in the Heat-Treatment of Cast Steel. C. T. Eddy, R. J. Marcotte and R. J. Smith. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1846: Metals Technology, 1945, vol. 12, Sept.). An investigation is reported the objects of which were: (1) To determine the S-curves for a 0·30%-carbon steel and three low-alloy chromium-nickel steels; (2) to ascertain whether the published S-curves for wrought steels can be used for cast steels of similar composition; and (3) to prepare S-curves for a number of low-alloy cast steels. For the four steels referred to in (1) it was found that there were only very slight differences between the curves for the wrought and the cast steels of similar composition.

Time-Temperature Relations in Tempering Steel. J. H. Hollomon and L. D. Jaffe. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1831: Metals Technology, 1945, vol. 12, Sept.). A study was made of the tempering of fully quenched steels differing only in carbon content, tempered over a wide range of time at temperatures from 65° F. to just below the critical point so as to establish a method of finding the combinations of temperature and time that produce the same hardness. The experimental details and the results obtained are given in Part I. and the method developed for the application of the results is described in Part II.

Coalescence of Cementite in the Products of the Decomposition of Austenite. H. Jolivet. (Revue de Métallurgie, Mémoires, 1943, vol. 40, Feb., pp 33–47; Mar., pp. 65–72). After discussing the principles on which the coalescence process in the solid solution is based, the author presents the results of his investigations and draws the following conclusions: The coalescence of cementite in the products of the decomposition of austenite takes place on heating the steel to below the Ac_1 point, and the finer the initial structure of the steel, the more rapid is the process; (2) the process of transformation is connected with the diffusion and dissolution of carbides; (3) the results of the heat-treatment can be estimated by microscopical examination, and the process corrected if necessary; (4) the determination of the size of the grains, by a method suggested by the author, considerably improved the results obtained.

Softening of Steel by the Isothermal Transformation of Austenite. F. F. Dodson. (Machine Shop Magazine, 1945, vol. 6, Nov., pp. 87–90). A comparison of ordinary annealing and isothermal annealing is presented. Each type of steel has its own optimum temperature for isothermal treatment, and to achieve the best results it is necessary to find the temperature at which the austenite begins to transform and the time taken for complete transformation. It is essential to allow sufficient time at the austenitizing temperature to bring the constituents to one homogeneous solid solution, and to control closely the temperature during the transformation period.

Heat-Treatment Study of Pearlitic Malleable Cast Iron. R. W. Lindsay and J. E. Atherton, jun. (American Foundryman, 1945, vol. 8, Sept., pp. 27–32). An investigation of the effects of two heat-treatments on the properties and microstructure of pearlitic malleable cast iron is reported. In the first, specimens were quenched in oil and tempered at subcritical temperatures for various lengths of time, and in the second, specimens were quenched to various subcritical temperatures at which they were held for different times. At a given tempering temperature

in the first treatment, the hardness of the martensitic structure decreased rapidly in the initial stages and then more gradually as the tempering time was lengthened. Isothermal transformation at from 1200° F. down to 1000° F. produced nodular pearlite which increased in fineness as the transformation temperature decreased. Isothermal transformation at below 1000° F. produced the acicular structure known as bainite.

Liquid Carburizing. J. O. Parker. (Australian Institute of Metals: Australasian Engineer, 1945, Oct., pp. 37–46). An account is given of the process of liquid-carburizing. The following aspects are discussed: Application and limitations of the process, selective carburizing, furnaces, containers, salts, bath control, nature of the process, and several methods of testing. The selection of suitable steels, recommended designs, effect of carbon content, case hardness and core toughness are also dealt with.

The Flame-Hardening of Different Qualities of Cast Iron. G. Kritzler and H. Killing. (Giesserei, 1942, vol. 29, Dec. 25, pp. 429–437). Several series of tests were made on bars 28 mm. in dia. of six qualities of cast iron to determine the effect of flame-hardening. The iron analyses ranged from 0·09% to 0·78% of combined carbon with 3·66% to 1·63% of silicon, and one contained 0·56% of molybdenum. A marked increase in the surface hardness of the ferritic cast iron was obtained by flame-hardening, and there was some increase in the core hardness on heating in a furnace and quenching which was not entirely accounted for by the heat-treatment alone. Chemical and metallographic examination showed that the matrix took up carbon from the graphite. A surface hardness as high as Brinell 550 was obtained with the pearlitic cast irons. All the hardened irons maintained their hardness on applying tempering treatments at 250–300° C. even after 200 hr. In the ferritic cast iron the combined carbon was increased from 0·09% to 1·03% by the furnace treatment, whereas with flame treatment for 25 sec. it increased to 0·7%.

Induction Hardening and Austenitizing Characteristics of Several Medium Carbon Steels. D. L. Martin and W. G. Van Note. (American Society for Metals, Oct., 1945, Preprint No. 17). The hardenability and transformation characteristics of a 0·50%-carbon steel, a 1·50%-manganese steel, a 3·50%-nickel steel, a chromium-nickel steel, and a manganese-silicon steel were studied to determine their suitability for induction-hardening. The steels with low Ac_1 and Ac_3 temperatures, little or no free ferrite, and medium to deep hardenability were ideal for induction-hardening.

Metallurgical Characteristics of Induction-Hardened Steel. J. W. Poynter. (American Society for Metals, Oct., 1945, Preprint No. 22). A comparison is made between the effects of induction-hardening and ordinary furnace-heating and quenching of bars of steel S.A.E. 4340 (a chromium-nickel-molybdenum steel). In the former case the depth of hardening is increased by increasing the heating time and reducing the power input, or by raising the surface temperature. The hardness depth is more easily controlled by the power input and time than by changes in the current frequency. What is often called "superhardness" in some induction-hardened steel seems to be related to the existence of internal stresses and is not caused by the induced currents.

Spark-Gap Converter as High-Frequency Power Source for Induction Heating. O. Weitmann. (Steel, 1945, vol. 117, Nov. 19, pp. 114, 130–132). The design of spark-gap converters and the circuits which have been developed for induction heating are discussed.

The Place of Radiant, Dielectric and Eddy-Current Heating in the Process Heating Field. L. J. C. Connell, O. W. Humphreys, and J. L. Rycroft. (Journal of the Institution of Electrical Engineers, 1945, vol. 92, Dec., pp. 464–465). Factors influencing the selection of one of the following methods of heating for a given industrial process are considered: (1) Conduction; (2) convection; (3) radiation; (4) dielectric heating; and (5) eddy-current heating.

Nitriding Engine Parts. W. Roth. (Steel, 1945, vol. 117, Nov. 19, pp. 123, 184–187). Plant and equipment for nitriding aero-engine cylinders, gears and crankshafts are described and illustrated.

A Mechanism of the Surface Decarburization of Steel. W. A. Pennington. (American Society for Metals, Oct., 1945, Preprint No. 30). A study was made of the decarburization of an ordinary carbon steel of eutectoid composition at temperatures in the 690–925° C. range. A mixture of hydrogen and water vapour was used to effect the decarburization. Micrographs are presented showing the progress of decarburization with time at the different temperatures. Distinct ferrite banding was found where the decarburization was carried out in the critical range (730–895° C.). There was banding below the lower critical point at 690° C., but it was not so distinct; there seemed to be no banding whatever at above the upper critical point. It is suggested that the carbon reaches the surface by diffusing in solid solution as either dissolved elementary carbon or iron carbide, regardless of whether the iron is in the austenitic or ferritic condition.

Preventing Surface Decarburization in Large Shaved Transmission Gears. E. G. de Coriolis. (Iron Age, 1945, vol. 156, Nov. 29, pp. 56–61). The methods and equipment used by a tank manufacturer for heat-treating large transmission gears after "shaving" are presented. The procedure described involves several innovations, notably the prevention of surface decarburization by the use of bell-type furnaces in which the gears are allowed to cool to less than 1000° F. before removal and subsequent hardening.

Heat-Treating Tank Parts in Buick Plant. (Industrial Heating, 1945, vol. 12, Aug., pp. 1292–1316). An illustrated description is given of a special heat-treating installation of the General Motors Corporation for the gas-carburizing, hardening, and tempering of parts for tank drives.

High-Speed Metal-Heating with Burners Radiant Ceramic. H. W. Smith, jun. (Engineering Materials and Processing Methods, 1945, vol. 22, Oct., pp. 1074–1078). A description is given of a ceramic gas burner for heat-treatment furnaces. The burner is cup-

shaped and recessed into the wall, and the design is such that the surface of the cup becomes incandescent so that the charge is heated by radiation.

Heat-Treatment. (Automobile Engineer, 1945, vol. 35, Nov., pp. 505–507). A description is given of a small vertical cylindrical forced-air-circulation furnace for heat-treating at temperatures up to about 700° C. It is equipped with two thermocouples recording the temperatures of the furnace and the charge on the same chart.

The Possibilities of Water-Quenching Certain Special Substitute Steels. G. Delbart and G. Wiltz. (Revue de Métallurgie, Mémoires, 1943, vol. 40, Nov., pp. 338–342). Tensile and impact specimens of chromium and chromium-vanadium steels, taken from 25-mm. billets, were subjected to water-quenching under different conditions, followed by tempering for 6 hr. at 650° C. The effect of quenching was observed, and tensile and impact tests were made as well as microscopical examination. It was found that slower or interrupted quenching seemed to be the best way to achieve a satisfactory hardening effect. The ageing symptoms were also investigated by means of Mesnager impact tests carried out after different treatments.

Quenching of Steel Balls and Rings. V. Paschkis. (American Society for Metals, Oct., 1945, Preprint No. 20). The temperature-time-volume relationships obtained when quenching steel spheres and cylindrical rings were investigated using the heat and mass flow analyser at Columbia University. General curves for spheres are presented which show the delaying effect of the heat of transformation in the 250–150° C. range. Data on the changes in the thermal conductivity and specific heat with temperature are also presented and discussed.

Production Cold-Treatment. A. J. T. Eyles. (Mechanical World, 1945, vol. 118, Nov. 16, pp. 543–545). The theory of the effect of cold-treatment in assisting the complete transformation of austenite to martensite in tool steels is explained, and procedures for treating various tools at −120° F. are given.

WELDING AND CUTTING

(Continued from p. 8 A)

Welding Mild Steel. H. O. Westendarp, jun. (Steel, 1945, vol. 117, Nov. 26, pp. 94–95, 123). Data are presented on the rates at which metal is deposited when welding mild steel at different current strengths using electrodes to the American Welding Society specifications, as well as on the properties of the welded joint obtained.

Brazing Alloy Tool Tips by Induction Heating. T. A. Vernor and E. F. Adams. (Steel Processing, 1945, vol. 31, Oct., pp. 652–653). An illustrated description is given of the equipment and process for utilizing induction heating for the brazing of tungsten-carbide tips to mild steel tools.

Aluminium Bonded to Steel or Cast Iron. M. G. Whitfield and V. Sheshunoff. (Engineering Materials and Processing Methods, 1945, vol. 22, Oct., pp. 1090–1096). Applications of the Al-Fin process of bonding aluminium to cast iron and steel are described with details of tests on the strength of the joint. The principal applications are fixing aluminium cooling fins on aero-engine cylinders, and making steel-backed aluminium bearings.

Modern Electric Weld Tube Mills. J. Lucas. (Iron and Steel Engineer, 1945, vol. 22, Oct., pp. 50–57). The development of machines for the continuous electric welding of strip into tubes of small and large diameter is described and illustrated.

Chemical Factors Affecting the Welding Properties of Stabilized 18-8 Stainless Steel. F. H. Page, jun. (Welding Journal, 1945, vol. 24, Oct., pp. 929–932). An investigation of the welding properties of 208 heats of stainless steel stabilized with titanium and 36 stabilized with columbium is reported. The results indicated that the cost of oxy-acetylene or atomic-hydrogen welding these steels may be materially decreased by slight changes in the chemical composition within the existing United States Army and Navy specifications.

The Spot Welding of NE 8715, NE 8630 and SAE 4340 in the 0·125-In. Thickness. W. F. Hess, W. D. Doty, and W. J. Childs (Welding Journal, 1945, vol. 24, Oct., pp. 521-S–530-S). This report describes work done for the War Production Board by the Rensselaer Polytechnic Institute in determining the optimum conditions for making tempered spot welds in three chromium-nickel-molybdenum steels. The tempering of the weld in the welding machine greatly improves the mechanical properties of the joint and makes it possible to spot-weld steels which are unweldable by conventional methods.

The Causes of Fissure Formation Tendency in Welded Chromium-Molybdenum Structural Steels. A. Antonioli. (Iron and Steel Institute, 1946, Translation Series, No. 261). This is an English translation of a paper which appeared in Stahl und Eisen, 1942, vol. 62, June 25, pp. 540–545 (*see* Journ. I. and S. I., 1943, No. I., p. 18 A).

Underwater Welding. J. R. Morrill. (Steel, 1945, vol. 117, Dec. 3, pp. 112–113, 153). Techniques for welding under water are described which were developed for the salvage of sunken vessels and repairing damaged ships, pipes and tanks.

Limitations in the X-Ray Testing of Welds. G. A. Homes and A. Portevin. (Revue de Métallurgie, Mémoires, 1943, vol. 40, Apr., pp. 97–107). Among the many factors which influence the X-ray testing of welds, the nature, size and location of defects are of great importance, and they form the subject discussed in this report. Artificially produced defects in welds are examined, and it is shown that certain internal defects cannot be detected by the X-rays alone; this method should be supplemented by

magnetic examination. This applies particularly to gas and slag inclusions the size of which is less than 5% of the thickness penetrated by the X-rays.

Removal and Repair of Steel Casting Defects. R. A. Pomfret. (American Foundryman, 1945, vol. 8, Sept., pp. 48–53). The techniques for cutting defective areas out of steel castings with the oxy-acetylene torch and of filling in with weld metal are described.

Some Developments in Oxy-Acetylene Applications. G. E. Bellew. (Iron and Steel Engineer, 1945, vol. 22, Oct., pp. 62–65). Applications of oxy-acetylene cutting in the steelworks are reviewed. These include deseaming billets, cutting risers from castings, and repairing ingot-stripper bits and tongs.

Recent Developments in Machine Flame Cutting. E. H. Witt. (Australian Welding Institute : Australasian Engineer, 1945, Oct., pp. 51–54). The present position of the machine flame-cutting process is outlined and different types of machines are described. Stack and heavy cutting, and the economic side of the process are dealt with.

MACHINING

(Continued from p. 9 A)

Milling Cast Iron with Carbide Cutters. (American Society of Mechanical Engineers : Machinery, 1945, vol. 67, Dec. 20, pp. 689–697). An extensive investigation is reported of the inter-relationship of tooth design, cutting speed, feed per tooth, type of cast iron, tool life, power required, and cost when milling cast iron with carbide-tipped cutters.

The Use of Carbide Tools. J. Jaquet. (Revue de Métallurgie, Mémoires, 1943, vol. 40, Aug., pp. 225–231). Rational methods and practical hints for the utilization of carbide tools are given, and the factors contributing to a higher production efficiency in comparison with high-speed steel tools are discussed.

Nomographs for Analysis of Metal-Cutting Processes. M. E. Merchant and N. Zlatin. (Mechanical Engineering, 1945, vol. 67, Nov., pp. 737–742). In this paper a number of charts and nomographs dealing with variables such as coefficient of friction, shear stress, frictional force, shear velocity, and cutting ratio, are presented for analysing all orthogonal cutting processes wherein a continuous chip is produced. (An orthogonal cutting process is one wherein the tool has a single straight cutting edge perpendicular to the direction of the tool motion relative to the work).

Detection, Causes and Prevention of Injury in Ground Surfaces. L. P. Tarasov. (American Society for Metals, Oct., 1945, Preprint No. 26). Information on injury to ground metal surfaces which is widely scattered in the literature is reviewed and supplemented by the author's work on stresses in ground surfaces. The subject is dealt with in the following order : (1) Methods of detecting cracks, stresses and burns in ground surfaces ; (2) cracks and crack patterns, including those occurring spontaneously during or after grinding and cracks revealed by etching ; (3) metallurgical factors which have repeatedly been shown to cause hardened steels to be sensitive to cracking during grinding ; (4) decreasing this sensitivity by improving the heat-treatment ; (5) the extent to which surface cracks may be injurious having regard to the subsequent application of the ground part ; and (6) methods by which stresses can be eliminated from surfaces after final grinding.

PROPERTIES AND TESTS

(Continued from pp. 11 A–14 A)

Cast Iron as a Material. E. Piwowarsky. (Giesserei, 1943, vol. 30, June, pp. 141–152). The properties of cast iron and its development as a constructional material are reviewed. Qualities of cast iron have been developed which can be rolled, forged, and pressure-die-cast. The Nipper rolling process for cast iron is referred to and brief details are given of a method of reinforcing iron castings with a very low carbon steel containing from 1% to 3% of silicon according to the thickness of the casting.

The Degree of Saturation and the Tensile Strength of Cast Iron. F. Bischof. (Giesserei, 1942, vol. 29, Dec. 25, pp. 437–439). The literature on formulæ for the degree of carbon saturation in cast iron and for the relation between the tensile strength and the saturation is reviewed, and difficulties in the determination of the tensile strength from the composition are discussed.

Gray Cast Iron. T. E. Barlow and C. H. Lorig. (American Foundryman, 1945, vol. 8, Sept., pp. 57–63). The relationships between the tensile strength, Brinell hardness and composition of grey cast iron are discussed.

Some Problems of the Metallic State. Sir Lawrence Bragg. (North-East Coast Institution of Engineers and Shipbuilders : Iron and Steel, 1945, vol. 18, Nov. 30, pp. 531–535). *See* p. 12 A.

Complex Properties in Metallurgy. A. Portevin. (Revue de Métallurgie, Mémoires, 1943, vol. 40, June, pp. 161–174). In addition to properties normally defined by scientific standard methods, there exist many other characteristic features of metals, such as, for example, flowability, forgeability, machinability, weldability, hardenability, corrodibility, etc., upon which the successful application of the metal depends. Taking into consideration the interdependence of these properties, which complicates production prob-

lems, the author stresses the necessity of examining them and recommends methods of testing.

Fracture of Metals Under Combined Stresses. D. J. McAdam, jun. (American Society for Metals, Oct., 1945, Preprint No. 29). The manner in which combined stresses contribute to brittle fracture is discussed. For a specific temperature, rate of deformation, and amount of previous plastic deformation, the resistance of a metal to fracture (the technical cohesive strength) may be represented by a curved surface in a diagram with the three principal stresses as co-ordinates. From two-dimensional diagrams based on the experiments of several investigators, three-dimensional diagrams have been derived to represent the technical cohesive strength of brittle and ductile metals.

The Fracture Stress of Steel. C. Zener. (Reviews of Modern Physics, 1945, vol. 17, Jan., pp. 20–26). A review of the results of research work at Watertown Arsenal on the fracture stress of steels is presented. The fracture-stress curve is that curve which gives for each strain the stress at which the material would fracture if further plastic deformation could be prevented; this curve and the flow-stress curve are discussed. If the fracture stress is to be controlled, the size of the microcracks within the steel must be controlled. The microcracks are in some way related to the interfaces between the various phases. The rôle of physicists in such investigations is considered.

The Theory of Strain and the Fracture of Solids with Special Reference to Deformation in Iron. A. Eichinger. (Archiv für das Eisenhüttenwesen, 1944, vol. 18, Sept.-Oct., pp. 73–88). The process of slip in metals is irreversible. As long as the intermediate conditions exist the rigid values applicable to the equilibria can be used only in a hypothetical sense. Because of the many ways in which the final condition is affected by the previous history, new factors (*e.g.*, energy and impulse distribution, laminar and turbulent flow) must be sought, and these can only be found by a deeper study of the constitution of matter. The conditions influencing the deformation of iron are critically examined and discussed.

The Notched-Bar Impact Strength of Cast Steel at Low Temperatures. H. Juretzek and W. Trommer. (Giesserei, 1943, vol. 30, Jan. 22, pp. 21–24). The value of the impact test as a criterion of the behaviour of low-alloy cast steels at temperatures down to −180° C. is discussed and the literature is reviewed. The results of impact tests at +20°, −80°, and −180° C. on steels alloyed with combinations of three of the following elements: Nickel, manganese, aluminium, chromium, silicon and vanadium, are presented.

Effects of Cold-Rolling on the True Stress-Strain Properties of a Low-Carbon Steel. F. J. Mehringer and C. W. MacGregor. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1849: Metals Technology, 1945, vol. 12, Sept.). The effects of cold-rolling on the true stress-strain properties of a low-carbon steel were investigated for total reductions up to 95%. Tension tests were made on the material after nearly complete ageing had taken place. The influence of heterogeneity and of the orientation of the specimen relative to the rolling direction were studied. The magnitude of the strain-ageing was also determined using true stress-strain relationships.

Stress Comparisons by Correlations with High Frequency Magnetic and Eddy Current Losses. P. E. Cavanagh. (American Society for Metals, Oct., 1945, Preprint No. 3). The possibility of using a high-frequency oscillator, the output of which is governed by magnetic and eddy-current losses, to make practical comparisons of stresses in metals is examined. Preliminary experiments are described which establish the fact that there is a correlation between high-frequency core losses and internal stresses. Examples are cited of the application of this method of testing to determine the optimum time of shot-peening required to produce compressive stresses at the surface of bars subjected to fatigue, and for predicting fatigue failures.

Relationships between the Deep-Drawing Properties of Autobody Sheet, Its Austenitic Grain Size and McQuaid-Ehn Carburizing Characteristics. K. J. B. Wolfe. (American Society for Metals, Oct., 1945, Preprint No. 27). An investigation of the causes of splitting during the intermediate forming operation in deep-drawing steel sheets is reported. Carburizing tests were carried out on samples from two batches, one giving consistently good results and the other consistently bad results in forming operations. The samples in the former case had fine austenite grains, whereas in the latter case they were coarse. The theory is advanced that this grain size depends on the distribution of carbide particles in the sheets and consequently on the carbide segregation in the ingot stage.

The Flowability of Metals and Alloys. P. Bastien. (Revue de Métallurgie, Mémoires, 1943, vol. 40, Dec. pp. 353–367). After discussing the laws of flowability of pure metals and alloys of binary and ternary systems the author gives the results of his experiments with a 0·25%-carbon open-hearth steel. The influence of temperature and of aluminium and silicon additions on the flowability of steel is described. The flowability is closely related to viscosity and superficial tension, and the testing of flowability, apart from being very useful in foundry practice, may serve as a helpful means of determining the solidification diagram for a given alloy.

Photo-Grid Techniques for Measuring Distortion. (Iron Age, 1945, vol. 156, Nov. 8, pp. 78–81). A technique, developed by an American corporation manufacturing aeroplanes, for producing grids on metal specimens is described. The grids are used for measuring distortion in tensile, punching, and drawing tests. Grids with 10 or 100 lines/in. can be produced.

Secondary Mechanical Properties. Internal Friction, Viscous Relaxation, Reactivity, Poisson's Ratio. P. Chevenard. (Revue de Métallurgie, Mémoires, 1943, vol. 40, Oct., pp. 289–300). The theory of resistance is based on the assumption that metals remain perfectly elastic below the limit of elasticity. The study of secondary mechanical properties, presented by the author, forms a valuable contribution to the recent research work which seeks to determine to what extent metals differ from a

perfectly elastic solid. The most sensitive parameter which defines this difference is the value of internal friction, measured by the Coulomb micro-pendulum. Poisson's ratio serves the same purpose. The Coulomb micro-pendulum and apparatus for measuring the Poisson's ratio are described in detail.

An Investigation of Different Qualities of Malleable Cast Iron with Reference to Their Torsional Fatigue Strength in the Smooth and Notched State, and Their Damping Capacity. J. Geiger. (Giesserei, 1943, vol. 30, Mar. 19, pp. 85–92). An earlier investigation of the damping capacity of cast iron (*see* Journ. I. and S. I., 1940, No. I., p. 205 A) has been extended to cover malleable cast iron. The results of damping measurements and torsional fatigue tests on three qualities of malleable cast iron using smooth, drilled and notched specimens are reported and discussed. The damping capacity was much higher than that of steel and the torsional fatigue strength was also higher. As the iron has good casting properties and is hardenable it is recommended for making aero-engine crankshafts, and tests on full-size crankshafts instead of on small specimens should be carried out.

Factors Affecting the Hardenability of Boron-Treated Steels. R. A. Grange and T. M. Garvey. (American Society for Metals, Oct., 1945, Preprint No. 10). The hardenabilities of four series of carbon steels (containing 0·40%, 0·52%, 0·63%, and 0·75% of carbon respectively), to each of which measured amounts of boron were added, are compared. The increase in hardenability due to boron is correlated with the boron content determined by a chemical method and with the amount of a characteristic constituent which appears in boron-treated steels after special heat-treatment. The increase in hardenability which results from the addition of the optimum amount of boron decreases with increasing carbon content; the inference is that boron will not increase the hardenability of hypereutectoid steel.

Determination of Most Efficient Alloy Combinations for Hardenability. H. E. Hostetter. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1905: Metals Technology, 1945, vol. 12, Sept.). Hardenability increment factors for chromium, nickel, and molybdenum are presented. These take into account the relative costs of these three elements when added to steel and the changes in the Grossmann "ideal critical diameter" of bar (*i.e.*, the size that will just harden to a 50%-martensite structure at the centre when given the fastest possible quench) with different additions. An explanation, with examples, is then given of a method of using the increment factors to determine the most economical composition which will result in a given hardenability.

Relationship between Hardenability and Percentage of Martensite in Some Low Alloy Steels. J. M. Hodge and M. A. Orehoski. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1800: Metals Technology, 1945, vol. 12, Sept.). The relationships between hardenability based on a 50%-martensite microstructure and hardenability on the basis of higher proportions of martensite have been studied by means of a metallographic survey of specimens subjected to the end-quench hardenability test. The results are presented by graphs which: (*a*) Correlate the hardenability on the 50% martensite basis with the hardenabilities on the basis of 99·5%, 95%, 90%, and 80% martensite structures; (*b*) correlate the calculated and actual hardenabilities on the basis of 99·9%, 95%, 90%, and 80% martensite structures; and (*c*) correlate the carbon content and the hardness of 99·9%, 95%, 90%, 80%, and 50% martensite structures.

The Influence of Titanium on the Hardenability of Steel. G. F. Comstock. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1904: Metals Technology, 1945, vol. 12, Sept.). The effect of titanium on the hardenability of steel depends on the heat-treatment, on whether the total titanium or the acid-soluble titanium is used as the basis for the calculations, and on the titanium/carbon ratio. Additions of up to 0·1% of titanium increase the hardenability effectively, but with amounts increasing above 0·2% the hardenability decreases. The hardenability factor for titanium may be based on the acid-soluble quantity present provided that account is taken of the fact that the insoluble titanium decreases the hardenability by rendering ineffective a proportional amount of carbon.

Microhardness Testing of Materials. V. E. Lysaght. (Engineering Materials and Processing Methods, 1945, vol. 22, Oct., pp. 1079–1084). Applications of the Knoop indenter are described. In particular, details are given of the Tukon instrument with which Knoop hardness numbers under loads in the range of 25 to 3600 g. can be obtained.

Hardness of Metals and Alloys. Its Measurement. L. Guillet. (Revue de Métallurgie, Mémoires, 1943, vol. 40, Sept., pp. 266–275). Hardness-testing methods, based on the principle of resistance against scratching and penetration, and of rebound after shock, are reviewed. Some theoretical explanations of hardness are also given. The hardness of metals seems to be connected with their atomic volume—the softer the metal, the greater its atomic volume. The value of hardness is considerably influenced by transformations which occur in solid solutions, such as precipitation, and also by diffusion, cold-work and recrystallization.

Ageing Phenomena in Cast Iron. E. Piwowarsky. (Giesserei, 1942, vol. 29, Oct. 16, p. 358). The ageing tendency of specimens from a special cast iron containing 3·6% of carbon, 2·6% of silicon and 0·75% of phosphorus was examined. They were heated for 20 min. at temperatures in the 250–550° C. range and quenched in water. Hardness determinations were made immediately and after intervals of 20, 40, 120 and 160 hr. The hardness of the specimens tested immediately after quenching increased markedly when the temperature of the treatment was increased from 350° to 450° C. After only 40 hr. a marked increase in hardness was noted, the maximum increase occurring after 160 hr.

A Method of Assessing Wear Due to Friction and Corrosion. P. M. Fisk. (Sheet Metal Industries, 1945, vol. 22, Dec., pp. 2138–2141, 2146). A description is given of a machine which was specially

made to study the wear and corrosion of hinges for doors and windows, and some results obtained with it are discussed.

Recent Studies on Metallic Wear By Sliding Friction. L. Decroix. (Revue de Métallurgie, Mémoires, 1943, vol. 40, Feb., pp. 48–57; Mar., pp. 73–84; Apr., pp. 114–122). The author presents a comprehensive review of recent research work on metallic wear by friction, and reports the results of his own investigations. The amount of wear depends primarily on the hardness and strength of the metal. Its mechanical properties are affected by the heat generated by friction, and the behaviour of the metal at high temperatures should be considered when the resistance to wear is tested. The process of friction causes chemical transformations in the worn layer by continuous reactions between gas, liquid and metal.

A Survey of the Theory of Ferromagnetism. J. H. Van Vleck. (Reviews of Modern Physics, 1945, vol. 17, Jan., pp. 27–47). There is similarity in many respects between the theory of ferromagnetism and the theory of valence. Both topics are readily tractable only for models corresponding to the limiting cases in which the electrons are treated respectively as tightly bound on individual atoms and as wandering freely from atom to atom without cognizance of each others' positions.

Effect of Small Stresses on Magnetic Properties. R. M. Bozorth and H. J. Williams. (Reviews of Modern Physics, 1945, vol. 17, Jan., pp. 72–80). Measurements were made of the changes of magnetization that occur in steel when very small stresses are applied and removed. Curves showing the changes in the magnetic properties are presented and discussed.

Mass Temperature Effects on Quenching 36 Per Cent. Cobalt Magnet Steel. B. Falk. (American Society for Metals, Oct., 1945, Preprint No. 6). The effect of mass on the magnetic properties of a 36%-cobalt steel are discussed and experiments are described the object of which was to determine the heat-treatments for this steel which would produce the optimum magnetic properties in the different commercial sizes. There is a different critical quenching temperature for different masses of the same steel which produces the maximum $BH_{max.}$ and this temperature (in Fahrenheit) could be found by the equation $T = 1605 + (912/F)$, where F is the ratio of surface area to volume.

Multi-Specimen Testing Equipment for Long-Time Creep Tests. A. Thum and K. Richard. (Iron and Steel Institute, 1946, Translation Series, No. 262). This is an English translation of a paper which appeared in Archiv für das Eisenhüttenwesen, 1943, vol. 17, July-Aug., pp. 29–33 (*see* Journ. I. and S. I., 1944, No. II., p. 133 A).

Investigation of a Type of Failure of 18-8 Stabilized Stainless Steel. W. Kahn, H. Oster and R. Wachtell. (American Society for Metals, Oct., 1945, Preprint No. 15). A report on an investigation of the failure of aero-engine exhaust manifolds of 18/8 stabilized stainless steel is presented. A type of failure was traced which was due to more carbon being precipitated in the steel than the stabilizing element could absorb.

Graphite in Cold-Rolled Subcritically Annealed Hypoeutectoid Steels. M. A. Hughes and J. G. Cutton. (American Society for Metals, Oct., 1945, Preprint No. 13). A study was made to determine the effect of (*a*) temperature, (*b*) residual alloying element, (*c*) cold-reduction, (*d*) mode of deoxidation, (*e*) full annealing before cold-reduction, and (*f*) carbon content, on the susceptibility of hypo-eutectoid steels to graphitization when cold-rolled and annealed at subcritical temperatures. Fine-grained steel strip with carbon in the 0·08–0·67% range was graphitized when cold-rolled and subcritically annealed. Hot-rolled silicon-manganese steel strip was also graphitized by this treatment. Data on the rate of solution of graphite in cold-rolled annealed strip at normal heat-treatment temperatures are presented.

Residual Elements and Gas Contents of Cast Iron. J. E. Hurst. (Staffordshire Iron and Steel Institute: Metallurgia, 1945, vol. 33, Nov., pp. 9–12: Foundry Trade Journal, 1945, vol. 77, Dec. 27, pp. 363–369). Examples are described of the marked effect which traces of residual elements or of gas in cast iron can have upon its physical properties. The examples relate to a high-silicon heat-resisting iron, an austenitic nickel cast-iron and a manganese iron for marine diesel-engine cylinders.

Gas Evolution from Cast Steel at Room Temperature. H. H. Johnson, L. H. Arner and H. A. Schwartz. (American Society for Metals, Oct., 1945, Preprint No. 14). The composition and rate of evolution of gas from freshly cast steel at or near room temperature was studied. The gases consisted mainly of carbon monoxide, nitrogen and hydrogen. The amount of the first two of these gases evolved from unit weight of electric-furnace medium-manganese steels differed very little from steel to steel, but the amount of hydrogen varied greatly.

Contribution to a Study of the Mechanical Properties of Special Substitute Steels in Relation to the Section of Thermally Treated Pieces. G. Delbart. (Revue de Métallurgie, Mémoires, 1943, vol. 40, Nov., pp. 321–337). The author presents the results of mechanical tests on the manganese, manganese-vanadium, manganese-chromium, and manganese-chromium-vanadium steels used as substitutes for high-alloy nickel-chromium steels. Test-pieces were taken in both longitudinal and transverse directions from billets of various dimensions (200, 100, 50 and 25 mm.) after heat-treatment. The study of the mechanical properties was supplemented by microscopical examination. An appropriate heat-treatment is suggested for each of the steels examined to obtain satisfactory results in pieces of large dimensions. It is also pointed out that a less homogeneous sorbitic structure, containing free ferrite after heat-treatment, may in certain cases improve their damping capacity.

Effect of Variations in Composition and Heat-Treatment on Some Properties of 4 to 6 Per Cent. Chromium Steel Containing Molybdenum and Titanium. G. F. Comstock. (American Society for Metals, Oct., 1945, PreprintNo. 4). With titanium/carbon ratios of between 3·5 and 5·5 in 5%-chromium/molybdenum/titanium steels, much better notched-bar impact strengths and high-temperature rupture strengths

are obtained than with higher titanium. About 1% of silicon in these steels gives superior resistance to oxidation and does not reduce the impact and high-temperature strengths if the manganese content is 0·75–1·00% and the steel is tempered at about 730° C. without annealing or normalizing.

Iron-Manganese Alloys. R. S. Dean, J. R. Long, T. R. Graham and R. G. Feustel. (American Society for Metals, Oct., 1945, Preprint No. 5). The effect of increasing amounts of manganese in the 1–7% range on the properties of cold-rolled steels was investigated. In the soft condition the tensile strength is increased by about 11,000 lb./sq. in. for each 1% of manganese. The elongation drops rapidly at first (by 5% for each 1% of manganese) and then more slowly at 7% manganese. Cold-working to 80% reduction in thickness increases the tensile strength to 192,600 lb./sq. in.; it produces an elongation of 2·7% on 2 in. and a hardness of Rockwell C 38. The effects of different heat-treatments are also discussed.

Effect of Nickel on Physical Properties and Thermal Characteristics of Some Cast Chromium-Molybdenum Steels. N. A. Ziegler and W. L. Meinhart. (American Society for Metals, Oct., 1945, Preprint No. 28). The effect of additions of up to 2% of nickel on the thermal and physical properties of steels containing 2·5–9·0% of chromium and 0·5–1·5% of molybdenum was investigated. The nickel improved the strength in all cases, without undue reduction of ductility and impact resistance. The effect was more pronounced with the 2·5–5·0% chromium steels than when the chromium content was 7–9%. The hardness developed on welding low-carbon nickel steels was less than that in high-carbon nickel-free steels of the same strength.

The Properties of Cast Manganese Steel and the Effect of Manganese on Cast Iron. F. Bischof. (Giesserei, 1942, vol. 29, Nov. 13, pp. 381–387). The properties of cast manganese steel, in particular steel with 10–12% of manganese, and the effects of manganese on the properties of cast iron were studied, special attention being given to contraction, expansion, thermal conductivity, wear-resistance, and work-hardening.

Stainless Steels for Turbine Blading. J. H. G. Monypenny. (Transactions of the Institute of Marine Engineers, 1945, vol. 57, Dec., pp. 129–140). *See* p. 14 A.

Selecting the Proper Heat-Resistant Steel. R. A. Lincoln. (Iron Age, 1945, vol. 156, Nov. 8, pp. 74–77). Data on the scaling-resistance and mechanical properties of the chromium-nickel heat-resisting steels are presented and discussed.

Ball Wear in Wet Grinding Mills. N. A. MacLeod. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1918 : Metals Technology, 1945, vol. 9, Nov.). Grinding tests were made in a small ball mill 10½ in. in dia. × 9 in. long, rotated at 72 r.p.m. Chill-cast, sand-cast and annealed steel balls were used to grind different types of ore and the wear was determined by the loss in weight. Siliceous ores and ores rich in pyrites caused a high rate of wear in wet-grinding, whilst the presence of kaolinitic material reduced the rate of wear. The grindability index of an ore is not necessarily an expression of its effect on ball consumption.

Selection of Die Steels for Cold Work. W. H. Wills. (Iron Age, 1945, vol. 156, Nov. 22, pp. 58–61). The structure and physical properties of the following four groups of die steels used for cold-work are described: (1) Water-hardening (carbon and carbon-vanadium) steels; (2) manganese and other low-alloy oil-hardening steels; (3) high-carbon high-chromium steels; and (4) air-hardening steels. Factors affecting their selection for special duties are discussed.

Electronic Measurement, Analysis and Inspection. (Machinery, 1945, vol. 67, Dec. 27, pp. 729–732). Industrial applications in which electronic devices are used for counting and detecting hidden defects, and controlling temperatures are described.

British *En* Alloy Specification. J. H. G. Monypenny. (Iron Age, 1945, vol. 156, Nov. 15, pp. 71–76). The British emergency specifications for steel are described and presented in the form of tables. The specifications were based on the physical and mechanical properties rather than on the chemical analysis.

METALLOGRAPHY AND CONSTITUTION

(Continued from p. 15 A)

Anti-Reflection Films for Metallographic Objectives. J. R. Benford. (American Society for Metals, Oct., 1945, Preprint No. 1). Experimental studies on improvements in the performance of metallographic microscopes due to anti-reflection films on the surface of the objective lens show that the improvement depends on the design of the objective and on the type of specimen. The improvements accomplished by the filming consist of a better image contrast and a shorter photographic exposure. Micrographs are submitted showing comparative performance between filmed and unfilmed objectives.

Electron Microscopic Investigations of Surface Structure. R. D. Heidenreich. (S.A.E. Journal, 1945, vol. 53, Oct., pp. 588–594). Examples of the condition of metal surfaces as revealed by the electron microscope are presented and discussed. The depth of focus of the electron microscope is of the order of 10 times that of the light microscope, so that structures with variations in elevation as great as several microns will yield micrographs that are uniformly in focus.

Defects in Steel and Their Detection. A. J. K. Honeyman and A. Fisher. (Journal of the West of Scotland Iron and Steel Institute, 1945, vol. 52, Part III., pp. 39–57). A brief account of defects in steel and methods of detecting them is given and illustrated by 38 figures and micrographs. Surface examination,

the deep-etch test, and the sulphur-print test are described. These reveal: (1) Pipe or axial segregation; (2) other types of ingot segregation; (3) reeds; (4) rokes; (5) shell; (6) overlaps; and (7) fishtail. Examples of each of these are described. Crack-detection tests are also described.

Detection of Deep Flaws in Metals by Supersonic Wave. D. O. Sproule. (Machinist, 1945, vol. 89, Dec. 8, pp. 2013–2016). A description is given of an instrument for transmitting supersonic waves through metal and receiving echoes from flaws within the metal. It is claimed that a $\frac{1}{32}$-in. dia. hole 4 ft. below the explored surface can easily be detected.

A Method of Revealing the Microstructure of Nickel Deposits on Steel. R. Stokes. (Metallurgia, 1945, vol. 33, Nov., pp. 21–22). A procedure for etching nickel deposits on steel which etched the nickel deeply with only slight attack on the steel and the interface is described.

Radiography and Shipyard Welding. R. Halmshaw. (Welding, 1945, vol. 13, Nov., pp. 451–455, 461). The principles of gamma-radiography are explained and a typical radon source 1 mm. in dia. × 2 mm. long in a glass bulb in a lead or tungsten-steel container is described. Some radiographs of shipyard welds are reproduced and discussed.

The Advantages and Limitations of Gamma-Ray Radiography on Small Steel Castings. R. H. Frank. (American Foundryman, 1945, vol. 8, Oct., pp. 50–61). A general and well-illustrated review of gamma-ray radiography as an inspection instrument and as a method of testing castings is given.

X-Ray Standards for Purchasing Castings. L. W. Ball. (Iron Age, 1945, vol. 156, Nov. 22, pp. 62–64).. A difficulty in drawing up acceptance specifications for castings is the unpredictability of internal discontinuities and the limited degree to which test-bars represent the quality of the corresponding cast parts. For the acceptance of castings, agreement can be furthered by examining radiographs and interpreting heterogeneity, if present, according to the importance of the area in which it is located.

Influence of Carbon Content upon the Transformation in 3 Per Cent. Chromium Steel. T. Lyman and A. R. Troiano. (American Society for Metals, Oct., 1945, Preprint No. 16). The effects of time and temperature on the transformations in seven 3%-chromium steels with carbon in the 0·08–1·28% range were investigated with the microscope, the dilatometer and by X-ray diffraction. Isothermal transformation diagrams for these steels are presented.

Transformation of Austenite in a Steel Containing 3 Per Cent. Chromium and 1 Per Cent. Carbon. E. P. Klier. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1855: Metals Technology, 1945, vol. 12, Sept.). An analysis was made of the transformation characteristics of a 1%-carbon, 3%-chromium steel by X-ray and metallographic methods. From the results obtained it is concluded that the bainite reaction in this steel is more complicated than previously reported by the author and T. Lyman for the low-carbon low-alloy steels.

Partition of Molybdenum in Hypereutectoid Iron-Carbon-Molybdenum Alloys. F. E. Bowman. (American Society for Metals, Oct., 1945, Preprint No. 2). The study of the partition of molybdenum between the ferrite and carbide resulting from the isothermal transformation of austenite as well as from tempering martensite is extended to hypo-eutectoid alloys of iron, carbon, and molybdenum. In these alloys the high concentration of molybdenum in the carbides formed isothermally at subcritical temperatures proves that molybdenum must diffuse during the eutectoid reaction. This diffusion of the molybdenum and the formation of the complex carbide $(Fe,Mo)_{23}C_6$ in alloys containing more than 0·50% of molybdenum provide an explanation for the retardation in the eutectoid reaction.

Critical Points of S.A.E. 4340 Steel as Determined by the Dilatometric Method. D. Niconoff. (American Society for Metals, Oct., 1945, Preprint No. 19). A dilatometer was employed to determine the critical points on heating a chromium-nickel-molybdenum steel. The position of the points depended upon the initial structure and the heating rate. The critical points on cooling depended on the time and temperature of soaking before cooling.

Isothermal Transformation of Austenite in One Per Cent. Carbon, High-Chromium Steels. T. Lyman and A. R. Troiano. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1801: Metals Technology, 1945, vol. 12, Sept.). Microscopical, X-ray and dilatometric methods have been employed in determining the isothermal transformation diagrams for four steels containing 1% of carbon with 3%, 4%, 6%, and 9% of chromium respectively.

Temper Brittleness of Steels. H. Jolivet and R. Chouteau. (Revue de Métallurgie, Mémoires, 1943, vol. 40, Jan., pp. 14–24). Slow cooling after tempering of certain steels produces temper brittleness, which manifests itself by a difference between the impact values obtained after rapid and slow cooling. There are two theories which endeavour to explain this: (1) The theory of precipitation of one phase in the austenitic zone, and (2) the theory of decomposition of carbides. The results of the investigations, presented by the authors lead to the conclusion that both theories remain hypothetical, as it was impossible with the microscope to produce evidence of the precipitation or decomposition of the carbides.

What Can Be Deduced from a Solidification Diagram? J. Luneau. (Revue de Métallurgie, Mémoires, 1942, vol. 39, July, pp. 218–223). The binary solidification diagram is discussed with particular reference to the three following problems: (1) The distribution of alloying elements between the liquid and solid phases at a given temperature; (2) the proportion in which both constituents participate in the solution when the temperature changes and the liquid phase passes into the solid phase; and (3) the graphical determination of the mass passing from the liquid to the solid phase when the temperature decreases.

CORROSION OF IRON AND STEEL

(Continued from p. 16 A)

Some Principles of Corrosion of Metals and Alloys. G. Chaudron. (Revue de Métallurgie, Mémoires, 1943, vol. 40, July, pp. 193–201). The symptoms of superficial and internal, or intergranular, corrosion are discussed in detail: The corrosion depends on a great many factors, which are divided by the author into four main groups: (1) The attacking reagent; (2) the metal or alloy examined; (3) the conditions of work; and (4) the duration of the corroding process.

Passivity of Metals. P. Lacombe. (Revue de Métallurgie, Mémoires, 1943, vol. 40, Sept., pp. 257–265). The author suggests the following definition of passivity: A metal is said to be passive when, during a chemical or electrochemical transformation, it loses a great deal of its power to emit ions, this emission being suppressed by interposition of a gaseous, solid or even liquid film. Various methods of producing and destroying passivity, and the factors on which it depends, are discussed. The existence of a film is the condition necessary for conferring passivity on a metal, but it is only effective when the film is very thin, continuous and adhesive, and of an amorphous structure.

Oxide-Metal Layers Formed on Commercial Iron-Silicon Alloys Exposed to High Temperatures. R. Ward. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1832: Metals Technology, 1945, vol. 12, Aug.). Structures of iron-silicon alloys containing from 0·70% to 5·8% of silicon were oxidized in air and in closed containers containing ferric oxide and mixtures of ferric oxide, iron powder and water at temperatures from 600° to 1200° C. The conclusions reached were: (1) At below 800° C. the diffusion of oxygen in alloys with less than 1·56% of silicon takes place preferentially along the grain boundaries; (2) at below 800° C. the diffusion in alloys with more than 2·55% of silicon appears to proceed by lattice diffusion; (3) when low-silicon alloys are oxidized at above 800° C. a layer of oxide particles is precipitated and this layer increases in depth with temperature and time; (4) with alloys containing more than 2·55% of silicon the character of the oxidation changes at 800° C. from a "pearlite" type to a structure similar to "Liesegang bands," and at above 1000° C. this in turn changes to a globular structure; (5) the Liesegang structure frequently comes off by spalling, whereas the pearlite scale is very often adherent; and (6) the character of the surface layers depends on the composition, temperature, rate of oxygen delivery, and, to a limited extent, time. Several micrographs of sections through the surface layers at 600 diameters are reproduced.

The Corrosive Action of Benzole Absorption Oils. C. M. Cawley and H. E. Newall. (Journal of the Society of Chemical Industry, 1945, vol. 64, Oct., pp. 285–290). This paper deals only with the corrosion which occurs in the stripping plant in the benzole absorption process. An account is given of a laboratory investigation with the object of determining the nature and origin of the substances causing corrosion. The main cause was found to be the action of ammonium thiocyanate formed by the reaction between ammonia and carbon disulphide in solution in the wash oil. The deposits of Prussian blue often found in benzole absorption plants are apparently formed by the reaction between ammonium thiocyanate and iron.

ANALYSIS

The Spectrographic Analysis of Austenitic Nickel-Chrome Steel. S. D. Steele and J. M. Johnston. (Journal of the Society of Chemical Industry, 1945, vol. 64, Oct., pp. 278–283). The object of the work recorded was to determine whether the range of spectrographic determinations could be extended to justify inclusion of both chromium and nickel in the spectro-analytical examination of steels. A satisfactory technique is described in detail whereby, from the viewpoint of production control analysis, determinations of chromium, nickel, silicon, manganese and aluminium can be made in chromium-nickel steels of the 18/8 type. The method incorporates the use of the flat-surface technique with a copper counter electrode. Plate calibration is obtained by the use of a rotating stepped sector wheel, the calibration curve being drawn by plotting densitometer readings against the logarithm of the relative intensities. The gap-width setting can be made within the permissible tolerance by the use of a feeler gauge. Assuming that the errors inherent in the method follow a symmetrical distribution, the accuracy which may be expected for determinations of chromium and nickel within the ranges 18% to 24% of chromium and 7% to 10% of nickel are $\pm 0{\cdot}47\%$ and $\pm 0{\cdot}20\%$ respectively.

The Occurrence of Calcium in Iron. (Bradley's Magazine, 1945, Dec., p. 7). A spectrographic technique for determining calcium in iron is described. From tests made on a number of irons it seems that all irons contain a small amount of calcium, the maximum range found being 0·005–0·010%. Large additions of calcium by means of calcium silicide to grey iron did not increase the calcium content of the final metal, which remained at 0·006%.

Modern Methods for the Quantitative Spectrochemical Analysis of Metal Alloys. R. Castro. (Revue de Métallurgie, Mémoires, 1942, vol. 39, Feb., pp. 54–60; Mar., pp. 84–96; Apr., pp. 112–122). A survey is presented of the spectrographic methods of analysing alloys with special reference to quantitative analysis. The author deals in separate sections

with the choice of rays, the analytical and photo-electric methods based on the measurement of differences in line intensity, the spectrochemical analysis of alloys in solution, and the elements detectable by spectrographic methods.

Modern Apparatus for Spectrochemical Qualitative and Quantitative Analysis of Alloys. R. Girschig. (Revue de Métallurgie, Mémoires, 1943, vol. 40, May, pp. 143–155; June, pp. 175–182; July, pp. 202–208; Aug., pp. 252–256). Apparatus and the principles on which the recent development of spectrochemical analysis is based are discussed. The precision of the spectrographic methods depends largely on the accuracy in reproducing the results of the measurements; this is needed to establish a true relation between the measured value and the chemical composition of the analysed substance. The evolution of spectrographic apparatus is marked by a wide application of automatic devices which enable the experimental conditions of the analysis to be recorded with great precision.

Colorimetric Determination of Molybdenum in Iron and Steel. M. Kapron and P. L. Hehman. (Industrial and Engineering Chemistry, Analytical Edition, 1945, vol. 17, Sept., pp. 573–576). A photometric method for determining molybdenum in ferrous metals is described. It makes use of water-soluble solvents of low volatility which produce a very stable molybdenum-thiocyanate-complex colour with no necessity for extraction.

Application of Colorimetry to the Analysis of Corrosion-Resistant Steels. L. G. Bricker, S. Weinberg, and K. L. Proctor. (Industrial and Engineering Chemistry, Analytical Edition, 1945, vol. 17, Oct., pp. 661–663). A photometric method for the determination of zinc in corrosion-resistant steels by the dithizone method is presented.

Acid Electric Furnace Carbon Control. C. Locke. (Iron Age, 1945, vol. 117, Oct. 18, pp. 69–71, 176 B). The literature on rapid methods of determining the carbon so as to be able to control heats in electric furnace steelmaking is reviewed, and the relationship between the carbon and ferrous-oxide contents of the bath are discussed.

The Estimation and Influence of Hydrogen in Weld Metal. E. C. Rollason and H. W. Mance. (Welding, 1945, vol. 13, Nov., pp. 436–440, 461). Methods of determining the hydrogen content of weld metal are described with data on the results obtained. The influence of hydrogen in welding and the embrittlement theory is discussed.

The Non-Destructive Testing of Material by the Spot Test. (Spot Tests for Nickel, Cobalt and Iron). R. Weihrich and F. Schwertner. (Archiv für das Eisenhüttenwesen, 1942, vol. 16, Aug., pp. 45–48). Rapid spot tests are described by which the presence of nickel, cobalt and iron in metal can be detected. In these tests a drop of a suitable etching agent is allowed to remain on the surface to be tested for a minute or two; the drop is then absorbed with an impregnated paper and a colour change indicates the presence of the element for which the test is made.

The Application of Drop Reactions on Steels. H. Fucke and M. Möhrle. (Archiv für das Eisenhüttenwesen, 1944, vol. 18, Sept.-Oct., pp. 47–56). Procedures are described for detecting the following elements in steel by means of drop reactions: Aluminium, chromium, cobalt, copper, manganese, molybdenum, nickel, titanium, vanadium and tungsten.

A Composite Method for the Determination of Silicon, Manganese and Phosphorus in Cast Iron and Steel. T. S. Harrison. (Analyst, 1945, vol. 70, Oct., pp. 362–365). A method is described for determining silicon gravimetrically, phosphorus absorptiometrically by the phosphovanadomolybdate technique, and manganese by absorptiometry or titration, all on a single sample of iron or steel.

Absorptiometric Methods of Analysis for 18/8 and Similar Steels. B. Rogers. (Metallurgia, 1945, vol. 33, Nov., pp. 13–16). Using the Hilger Spekker photo-electric absorptiometer, E. J. Vaughan's work has been adapted in the direction of routine methods for the determination of chromium, nickel, manganese, and molybdenum on one sample weight for 18/8 stainless and similar steels containing 12–25% of chromium and 6–14% of nickel.

The Determination of Arsenic in Ferro-Alloys with Special Reference to Ferrotungsten. R. Weihrich and J. Haas. (Archiv für das Eisenhüttenwesen, 1942, vol. 16, Oct., pp. 129–130). When ferrotungsten is dissolved in a mixture of nitric and hydrofluoric acids some of the arsenic is volatilized, but if the sample is first melted with sodium peroxide no arsenic is lost and more accurate arsenic determinations will be made. Details of both procedures are given.

The Determination of Traces of Sulphur Dioxide with Special Reference to the Determination of Sulphur in Ferro-Alloys. G. Ingram. (Analyst, 1945, vol. 70, Nov., pp. 423–426). The method described for the determination of trace amounts of sulphur dioxide was developed to make possible the satisfactory estimation of sulphur in ferro-alloys and organic compounds using the minimum of material, *e.g.*, 10–50 mg. Amounts of sulphur dioxide exceeding 2 μg. can be determined.

New Type Furnace for the Determination of Ash from Coal and Coke. C. W. G. Ockelford. (Fuel in Science and Practice, 1945, vol. 24, Nov.-Dec., pp. 151–157). Developments since 1920 in the design of furnaces used for determining the ash in coal by the Fuel Research Organization method are described and details are given of two modern electric furnaces for temperatures up to 500° and 775° C. respectively.

BOOK NOTICES

(Continued from p. 16 A)

BELCHER, R., and A. L. GODBERT. "*Semi-Micro Quantitative Organic Analysis.*" 8vo, pp. viii + 168. Illustrated. London, 1945: Longmans, Green and Co., Ltd. (Price 10*s.* 6*d.*)

Although semimicro-analysis is a distinct subject in itself, it holds an important intermediate position between macro- and micro-methods, having the dual advantages of requiring less manipulative skill than micro-analysis and of effecting savings in samples, reagents and time, as compared with macro-methods. The purpose of the present book, as pointed out in the preface, is to provide a complete course in the more common methods of this type of analysis, the methods given, it is claimed, being simple and accurate.

The book is divided into four parts. The first part is composed of an introduction to the subject and a general survey of apparatus and technique. This section includes an extraordinarily good chapter "The Balance and Methods of Weighing" from which even well-informed chemists can learn quite a lot about the use and care of the balance. The second part of the book is devoted to the determination of the elements carbon, hydrogen, nitrogen, sulphur, the halogens, phosphorus and arsenic. The first chapter of this section deals with the determination of moisture, ash and metals. Next, there is a short section describing the determination of three of the more important organic radicals—the carboxyl group, the methoxyl group and the acetyl group. The final section is concerned with physico-chemical determinations, limited to the following four properties: density, melting point, boiling point and molecular weight.

In addition to these main sections there are two useful appendices, one being a thorough consideration of the various methods available for the purification of small amounts of material; the other is relatively short and describes the preparation and standardization of seven standard volumetric solutions. There is also appended a bibliography of 46 references.

The book is attractively produced, clearly arranged, and includes 42 diagrams of apparatus drawn with the utmost clarity. It might be mentioned that a more extensive index would have been welcomed. W. W. STEVENSON.

BRITISH NON-FERROUS METALS RESEARCH ASSOCIATION. "*Collected Papers on Metallurgical Analysis by the Spectrograph.*" Edited by D. M. Smith. 8vo, pp. xi + 162. Illustrated. London, 1945: The Association. (Price 21*s.*)

This book, compiled by specialists for the use of specialists, emphasizes the encroachment of purely physical methods upon the classical technique of metallurgical analysis.

Some interest attaches to the Foreword, in which the advantages of co-operative research work become self-evident, especially where association is possible with manufacturers of ancillary materials—in this case, manufacturers of photographic materials.

Much sympathy will be felt for the plea expressed in the Preface for a greater uniformity in technical nomenclature. When experts are not agreed upon the correct description of their methods, assimilation of such progress by technicians in other fields undoubtedly becomes more difficult.

Only non-ferrous metals and their alloys are dealt with in these papers, and the methods recommended appear with precise and detailed items of procedure which are obviously the fruits of close study and considerable experience; one may instance the care taken in preparation of the sample under test (p. 83).

Readers who are mathematically-minded will appreciate frequent allusions to "standard error" and "standard deviation"—evidence of the author's leaning to statistical methods in the assessment of results.

J. FERDINAND KAYSER.

HEDLEY, W. H. "*The Basis of Sheet Metal Drafting.*" 8vo, pp. viii + 118. Illustrated. London, 1945: Longmans, Green and Co., Ltd. (Price 6*s.*)

The purpose of this book is to give a firm foundation in the setting out of patterns, which forms an important part in the technical education of those connected with sheet metal work. Students should find the application of calculations to the development of patterns interesting and valuable.

Although written primarily for students taking the first year senior course of the Northern Counties Technical Council and other similar examining bodies, the book should have a special appeal to the practical metal plate worker; and all interested in handicraft will find it a useful text-book. The contents are as follows: (1) Instruments and Drawing Equipment; (2) Geometrical Constructions using Rule and Compasses; (3) Geometrical Constructions Involving the Use of Drawing-Board, Tee-square, Set Squares, etc.; (4) Orthographic Projection; (5) True Shapes and Sections; (6) Isometric Projection; (7) The Straight Line and Plane; (8) Developments; (9) Triangulation; (10) Intersection of Surfaces and Developments; (11) Developing without Projecting.

VOGEL, A. I. "*A Text Book of Qualitative Chemical Analysis, including Semimicro Qualitative Analysis.*" Third Ed., 8vo, pp. xii + 578. Illustrated. London, 1945: Longmans, Green and Co., Ltd. (Price 12*s.* 6*d.*)

The two previous editions of this book are so well known that the main purpose of this review must be to indicate the alterations and additions that have been made in this third edition. The book is intended to provide a complete course in inorganic qualitative analysis to degree standard, as well as a work of reference for practical analytical chemists.

The first chapter, which is an excellent one, on the theory of qualitative analysis remains unchanged, whilst only a few minor modifications, in the microchemical section, have been made to the second chapter on "Analytical Operations," which is a well-designed introduction to experimental technique. The next two chapters, which deal in detail with the reactions of metal ions and acid radicals respectively, remain substantially unaltered. Chapter V., describing the "Systematic Analysis of Inorganic Substances," has been considerably extended to give a more complete account of the examination for acid radicals. Chapter VI. is concerned with the modifications in analytical methods required in the presence of organic acids, silicates, borates, fluorides and phosphates. The reactions of the so-called rarer elements are dealt with in Chapter VII.

Here follow two completely new chapters. Chapter VIII., for elementary students, describes a systematic qualitative analysis of inorganic substances. In view of the number of elementary books on the subject, this chapter does not appear to be a necessary addition to the present volume.

Chapter IX. is devoted to semimicro-methods and is a most welcome addition. The first 23 pages are devoted to an introduction to the new experimental technique required, whilst the remaining 33 pages provide a systematic scheme for semimicro qualitative analysis.

The Appendix contains such information as is to be expected in a book on this subject, together with some model analyses for the guidance of students.

One minor criticism concerns the method of cross-reference, by section numbers instead of page numbers, which is not conducive to quick reference.

W. W. STEVENSON.

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MINERAL RESOURCES

(Continued from p. 1 A)

The Coking Coal Resources in the Yorkshire, Nottinghamshire and Derbyshire Coalfield. R. A. Mott. (Transactions of the Institution of Mining Engineers, 1945, vol. 104, June, pp. 446–468).

The Lancashire Coalfield. N. Simpkin. (Institute of Fuel War Time Bulletin, 1945, Dec., pp. 54–63). A general description and a survey of the history of the Lancashire coalfield are given; these cover its geological structure, coal seams, lower coal measures, output, and types and analyses of coal.

The Grand Lake Coal Field. N.B. W. J. Wright and C. S. Clements. (Transactions of the Canadian Institute of Mining and Metallurgy, 1945, vol. 48, Dec., pp. 747–754). A description is given of the geology and present state of development of the Grand Lake coalfield of central New Brunswick from which 368,544 short tons of bituminous coal were mined in 1943.

The Production of Both Coking and Steam Coals from the Witbank Coalfield. L. A. Bushell, S. B. Gass, F. W. Quass, and R. F. J. Teichmann. (Journal of the Chemical, Metallurgical and Mining Society of South Africa, 1945, vol. 46, July-Aug., pp. 45–59). Results of some float-and-sink analyses and coking tests on coking and steam coals from the Witbank coalfield in South Africa are reported and discussed. It is shown that the reserves of coking coal are small, but by the immediate introduction of blending and other economic measures the life of the reserves could be increased to about 163 years.

Gold-Coast Manganese. (Iron and Steel, 1945, vol. 18, Nov. 30, pp. 536–539). A brief and well-illustrated description is given of the open-cast workings of the manganese ore deposits at Nsuta in the Western Province of the Gold Coast. There are a washing plant and mechanical units to crush and grade the ore.

ORES—MINING AND TREATMENT

Process Measures in the Sintering of Iron Ores. K. Kintzinger. (Stahl und Eisen, 1943, vol. 63, June 10, pp. 453–456). An account is given of laboratory and full-scale investigations of methods of improving the unsatisfactory output of German sintering plants. Increasing the suction in the range of 600–1600 mm. of water was tried, and the optimum suction was found to be at about 1200 mm. Improvements were also obtained by mixing ground burnt lime with the charge.

Principles of Iron Ore Beneficiation and Their Effects on Blast-Furnace Operation. C. E. Agnew. (Steel, 1945, vol. 117, Nov. 26, pp. 124–126, 144–158; Dec. 3, pp. 134, 170–186; Dec. 10, pp. 122–126, 144; Dec. 17, pp. 125–128, 164–168). The thermal principles of blast-furnace operation as applied to the use of sinter were dealt with previously (*see* Journ. I. and S.I., 1944, No. I., p. 42 A). In this paper the beneficiation of ores by sintering, and by making nodules, briquettes and pellets is described. The properties of the products are discussed. The effect of the chemical composition of the slag on the thermal requirements of the hearth and bosh are also dealt with.

New Magnetic Concentrator Treats Mt. Hope Iron Ore. H. Davenport. (Engineering and Mining Journal, 1945, vol. 146, Sept., pp. 85–89). An illustrated description is given of the plant for concentrating the iron ore at the Mount Hope Mine, Wharton, New Jersey. In view of the heavy demand for high-grade lump magnetite, the new concentrator produces approximately two-thirds of the concentrates in the lump-ore plant and one-third by wet magnetic concentration in the fine-ore plant.

FUEL

(Continued from p. 17 A)

Theoretical Limiting Efficiency of Various Fuels in the Open Hearth. B. M. Larsen and C. Siddall. (Iron and Steel Engineer, 1945, vol. 22, Dec., pp. 76–95). A detailed theoretical consideration governing the selection and use of fuels for the open-hearth furnace is presented. Thermal data and examples of calculations are also given.

Blast-Furnace Gas and Oxygen-Enriched Air for the Open-Hearth Furnace. G. Bulle. (Zeitschrift des Vereines deutscher Ingenieure, 1940, vol. 84, Sept. 28, pp. 740–741). As the cost of producing oxygen has been reduced considerably by the Linde-Fränkl process, tests were made to determine the applicability of mixtures (*a*) of blast-furnace gas and oxygen, and (*b*) of blast-furnace gas and oxygen-enriched air for steelmaking in the open-hearth furnace. A horizontal cylindrical furnace 2 m. long with a capacity of 100 kg. of steel was used. The charges were easily melted with either of the above mixtures (the fuel gases being heated in recuperators), and the data obtained were sufficiently encouraging for a larger experimental plant to be built.

Fuel Problems in the Swedish Iron and Steel Industry. M. Tigerschiöld. (Iron and Steel Institute, 1946, this Journal, Section I.). The evolution of the Swedish iron and steel industry and the mineralogical, historical, and economic factors affecting that evolution are briefly discussed. The occupation of Norway by the Germans in 1940 completely isolated Sweden

from British and American sources of imports; with little or no suitable native resources of coal, coke, petrol or fuel-oil, the position of Swedish industry in regard to fuel supplies was critical. Fuel economy became a paramount necessity and industry went over to the use of substitutes, such material as wood, charcoal, and pressed-straw bales being used as metallurgical fuels. Equipment and methods were altered and new equipment and methods were devised to meet the situation, with striking results.

Fuel Utilization in Iron and Steel Works. N. H. Turner and F. A. Gray. (Iron and Steel Institute, 1946, this Journal, Section I.). Problems met with in current fuel-utilization practice at integrated and non-integrated iron and steel works are briefly discussed. These problems are chiefly concerned with blast-furnace gas production, cleaning, and distribution; some remarks as to the effect on fuel utilization of possible future modifications in blast-furnace practice have been added. The application of by-product gases to steelworks uses and the general question of heat conservation are also considered. A comparison is made of the operating results at the melting shops and mills of the Appleby and the Templeborough Works. The paper concludes with some indication of what has been achieved, in respect of fuel economy, at certain subsidiary plants of The United Steel Companies, Ltd.

Operating Results of Dry-Cooling Plants. (Coke and Smokeless Fuel Age, 1946, vol. 8, Jan., pp. 10–11). The dry cooling of coke has been developed in Switzerland on a considerable scale. In this short article two tables are presented by the Swiss Association of Gas Engineers giving the characteristics and operating results of twenty dry-cooling plants.

Upgrading Coke-Oven Gas. G. Cellan-Jones. (Coke and Smokeless Fuel Age, 1946, vol. 8, Jan., pp. 8–9). A method of increasing the benzole yield from coke-oven gas by introducing stripped gas into the oven crown is described.

Gaseous Fuels in Britain. D. D. Howat. (Chemical Age, 1946, vol. 54, Jan. 26, pp. 107–111; Feb. 2, pp. 133–137). The efficient utilization of town gas, coke-oven gas, blast-furnace gas, and producer gas in Great Britain is discussed.

Operational Experience with a Gas-Producer Plant with Subsequent Electric Gas Cleaning. E. Pechal. (Iron and Steel Institute, 1946, Translation Series, No. 265). This is an English translation of a paper which appeared in Stahl und Eisen, 1943, vol. 63, Nov. 4, pp. 808–812 (*see* Journ. I. and S.I., 1945, No. II., p. 178 A).

PRODUCTION OF IRON

(Continued from p. 18 A)

Designs of Withdrawal Lugs for Blast-Furnace Cooling Plates. J. K. Reay and R. P. Towndrow. (Iron and Coal Trades Review, 1946, vol. 152, Jan. 4, pp. 16, 15). A design is published of a lug, cast integrally with copper cooling plates for blast-furnaces, which is capable of withstanding a direct pull of about 25 tons.

Smelting Iron Ore in the Electric Low-Shaft Furnace. M. A. Kassen. (Schweizer Archiv, 1945, vol. 11, Mar., pp. 87–93). The control of the Tysland-Hole electric furnace for making pig-iron is discussed. The method of controlling the silicon by adjusting the coke in the burden, which is applied in blast-furnace operation cannot be applied in the electric process, but silicon control by adjustment of the amperage/voltage ratio is possible. The method is dealt with fully in this paper.

The Höganäs Sponge Iron Process. S. Eketorp. (Svenska Teknologförening: Jernkontorets Annaler, 1945, vol. 129, No. 12, pp. 703–721). (In Swedish). A description is given of the history and present stage of development of the Höganäs process of making sponge iron. It consists of charging ceramic pots about 5 ft. high, made of a local fireclay, with flat briquettes of a concentrate containing 71·5% of iron interspersed with layers of carbon prepared by mixing Höganäs coal with coke breeze; the pots are charged in batches in a long pit furnace where they are heated to about 1200° C., the total time in the furnace being about 12 days. The pots are emptied by manual labour and put on a conveyor which takes them to tumbling machines where the ash and dirt are removed. The furnace is large enough to take 35,000 pots. The process has been in operation since 1911, and production is now at the rate of about 20,000 tons/annum. The product contains 97·2% of iron, 0·009% of sulphur and 0·011% of phosphorus. As produced, the sponge-iron blocks have a density of 2·0, but if they are compressed into briquettes the density is about 3·5.

A brief account is given of the use of this sponge iron in steelmaking. In a trial at Surahammar a charge of 100% briquettes was melted in a basic-lined high-frequency furnace and the wear on the lining was not abnormal. In acid-lined furnaces it was not possible to charge more than 20% briquettes until the experiment of making the sponge iron with a certain amount of silicon was tried, which reduced the lining wear to normal.

The Krupp-Renn Process. (Metal Treatment, 1945, vol. 12, Winter Issue, pp. 263–271). Brief descriptions of the Walz and Renn rotary furnaces and the results of trial smelting runs in which the Krupp-Renn process is used in three different applications are given. The first application is the production of "pellets" for charging directly into the open-hearth furnace; the second is the production of pellets for the blast-furnace; and the third is the production of non-ferrous metals such as nickel, copper, silver, and cobalt, in each case with iron as well.

Powder Metallurgy of Iron. G. H. S. Price. (Metal Treatment, 1945, vol. 12, Winter Issue, pp. 275–286, 274). A comprehensive review of the literature on the powder metallurgy of iron is presented.

FOUNDRY PRACTICE

(Continued from pp. 18 A–21 A)

Cast Iron Foundry Practice. United States Wartime Developments. J. S. Vanick. (Association Technique de Fonderie : American Foundryman, 1945, vol. 8, Nov., pp. 57–67). The progress made in American foundries in the period 1940 to 1945 to increase production is described and illustrated. Special reference is made to the turntable system of moulding and casting.

Some Thoughts on Foundry Management. J. R. Roxburgh. (Institute of British Foundrymen : Foundry Trade Journal, 1946, vol. 78, Jan. 24, pp. 75–81). Principles of good works management are expounded, and it is stressed that executives should possess the greatest possible degree of practical experience in the foundry. Some of the problems appertaining to foundry management, *e.g.*, time and motion study, production control, costing, and working conditions, are separately dealt with.

Cast Steel. (Iron and Steel, 1945, vol. 18, Dec., p. 628). Thorough deoxidation with a view to minimizing porosity, internal tears, low ductility, and poor surfaces, is easily accomplished by the use of silicomanganese and calcium alloys. To assure consistently good results a uniform furnace procedure is necessary ; such a procedure is outlined.

Malleable Sand Control in a Large Mechanized Foundry. J. J. Clark. (American Foundryman, 1945, vol. 8, Nov., pp. 49–56). A description is given of the equipment and the tests designed to enable good moulding sand and cores to be furnished to the foundry. The importance of proper maintenance and operation of the equipment is stressed.

Sand Supply to Moulding Machines from Overhead Hoppers. N. C. Blythe. (Institute of British Foundrymen : Foundry Trade Journal, 1946, vol. 78, Jan. 3, pp. 9–11). The design of suitable hoppers which give satisfactory service under the severe conditions of the moulding shop is discussed. A type of hopper is suggested which provides an operator with sand every time he wants it. A distribution conveyor to take care of the hopper's failings in capacity is also described.

The Application of Ethyl Silicate to Foundry Practice. C. Shaw. (Institute of Britsh Foundrymen and the Institute of Metals : Foundry Trade Journal, 1946, vol. 78, Jan. 10, pp. 31–33). Refractory bodies can be produced by using silicon esters as a binder for refractory aggregates. When fully hydrolysed, the addition of an alkali to the liquid ethyl silicate will so shorten the time of gelatinization that, by varying the amount of alkali, the cement will set after a predetermined time. Certain substances, the nature of which is not disclosed, promote the formation of strong gels and sound moulds, the hydrolysis going on in the mould. The setting time may be varied between 5 min. and 5 hr., according to the amount of water added and the concentration of the agent. Many uses for this method in the foundry industry are recommended, and it is suggested that many moulding problems can be solved by taking advantage of the controlled setting time, the dimensional stability during setting and baking, and the high mechanical strength. Moulding from wax patterns and the " semi-dry piece mould " are also described.

Patternmaking for Small Batch Machine-Moulded Castings. A. G. Squires. (Institute of British Foundrymen : Foundry Trade Journal, 1946, vol. 78, Jan. 10, pp. 35–37). A method of preparing moulds for small batches of similar castings where half of the pattern will fall below the face of the moulding plate is described. It comprises a pair of pattern plates with six plaster patterns matched and produced from a single master pattern.

Pattern Redesign for Increased Production. E. C. Moorhead. (American Foundryman, 1945, vol. 8, Nov., pp. 26–28). In many foundries where castings are made from an original pattern over long periods production could in many cases be greatly increased by minor changes in pattern design and equipment without necessarily changing the casting itself. Recommendations on redesigning patterns and simplifying moulding are made.

Quantity Floor Moulding—Manufacture of Medium Bore Flanged Return Bends. I. H. List. (Iron and Steel, 1945, vol. 18, Dec., p. 646). Sketches show the pattern for medium-bore flanged return bends and a cross-section through the mould. Details of the mould and moulding operation are given.

The Influence of Radiation within Molding Sand on the Freezing Rate of Metal. H. A. Schwartz. (American Foundryman, 1945, vol. 8, Nov., pp. 45–48). A mathematical and theoretical treatment of the heat transfer within moulding sand is presented. The relationship between the amount of metal which freezes in a sand mould and time cannot be satisfactorily explained if it is assumed that the thermal conductivity of sand is independent of temperature. In this paper calculations are put forward which indicate that the transfer of heat through sand takes place partly by conduction and partly by radiation.

Steel Castings Repair Methods. J. F. Cotton. (American Foundryman, 1945, vol. 8, Nov., pp. 29–34). Various methods of examining and removing defects in steel castings are described. Guidance is given for deciding whether repair is necessary or desirable and on how the repair should be effected.

Flame-Gouging of Steel Castings. A. E. Blake, jun. (Engineering Materials and Processing Methods, 1945, vol. 22, Nov., pp. 1428–1431). The technique for the removal of defective areas from steel castings by flame-gouging with a burner using town gas and oxygen is described.

Notes on Inspection. R. J. Hart. (Institute of British Foundrymen : Foundry Trade Journal, 1946, vol. 78, Jan. 3, pp. 3–8, 21). The methods employed in routine inspection, both of pattern equipment and castings (mainly steel castings), are outlined and illustrated. Apart from the object of inspection, the principal points discussed are : Inspection of patterns and samples, bulk inspection, physical testing, and delegated inspection.

PRODUCTION OF STEEL

(Continued from pp. 21 A–22 A)

Soviet Iron and Steel. Main Trends of Developments in the New Five Year Plan. I. Bardin. (Iron and Steel, 1945, vol. 18, Dec., pp. 657–658). The immediate aims and objects of the Russian steel industry are outlined, and it is implied that the U.S.S.R. aims at self-sufficiency in ferrous materials. Principles of managing iron and steel plants are put forward, and it is emphasized that obsolete plants should be discarded, although they may be in a good state of repair. With a view to establishing a minimum number of types of blast-furnace, standardization of plant and equipment generally is envisaged.

The German Steel Industry. T. W. Lippert. (Iron Age, 1945, vol. 156, Dec. 20, pp. 54–59). A detailed account is given of plant, processes, labour employed and the working conditions in German steelworks during the last war. High production was ensured by imported labour and some technical innovations. An all-basic open-hearth furnace practice was used. Among the technical improvements the number of cranes and charging machines was surprisingly large; and the health service for German labour was of a high standard although safety precautions were of a low order.

General Review of German Metallurgical Practices. J. H. Frye. (Metal Progress, 1946, vol. 49, Jan., pp. 76–86). In 1945 the author examined many of the German metallurgical plants and processes and in this review he reports on the state in which he found them and on the changes and developments which had taken place during the war years. He compares the efficiency of the processes and products with those obtaining in the United States.

Advancements in the Metallurgical Sphere in the United States during the War. A. Grønninsæter. (Norsk Metallurgisk Selskap: Jernkontorets Annaler, 1945, vol. 129, No. 12, pp. 683–702). (In Norwegian). This paper, which has appeared in Tidsskrift for Kjemi og Bergvesen, reviews the technical advancements in the production of iron and steel, sulphur, manganese, copper, lead, zinc, chromium, aluminium, magnesium and the ferro-alloys, which have taken place in the United States in the period 1940–1945. Methods of concentrating ores are also dealt with.

Electric Power in a Steel Plant. R. W. Graham. (Electrical Engineering, 1945, vol. 64, Dec., pp. 816–818). A brief description is given of the system adopted by the Bethlehem Steel Company for the distribution of electric power at their Lackawanna plant.

Proceedings of the Twenty-Eighth National Conference on Open-Hearth Steel. (American Institute of Mining and Metallurgical Engineers, Open-Hearth Conference, Apr.-May, 1945). In order to economize in transport and hotel accommodation it was decided to hold four local conferences at Philadelphia (Apr. 20), Chicago (Apr. 27), St. Louis (May 4), and Pittsburgh (May 18), respectively, instead of one large meeting at Chicago as originally intended.

Two special papers are given, namely "Melt Control, Its Evolution and the Effect of a Current Control Method on Basic Open-Hearth Operations" for which the author, A. M. Kroner, received the McKune award, and "Operation of an Experimental Open-Hearth Furnace" by H. K. Work and W. R. Webb (*see* Journ. I. and S.I., 1945, No. II., p. 136 A). Kroner's paper describes a method of calculating the hot-metal additions for open-hearth furnaces which takes the following factors into account: (*a*) The percentage of iron in the metallic charge; (*b*) the amounts of scrap, cold iron, limestone and ore charged; (*c*) the silicon content of the hot metal; and (*d*) the time elapsing between starting to charge cold metal and finishing the hot-metal addition.

The Pittsburgh meeting, which dealt with refractories for open-hearth furnaces, is next reported. This began with a report by H. M. Griffith on the operation, and condition after 500 heats, of an all-basic-lined 180-ton open-hearth furnace of the Steel Company of Canada. Experiences with all-basic-lined open-hearth furnaces were related by H. S. Frank and A. E. Reinhard, and R. B. Sosman presented a paper on "The Outlook for an All-Basic Open-Hearth Furnace." After a brief discussion of silica-brick roofs, M. A. Fay dealt with a new theory on the mechanism of wear of these roofs which is that the wear of roof bricks is caused by liquid flowing from a considerable distance within the brick. The merits of bonded and of ribbed roofs, and mullite refractories were discussed, and W. S. Debenham presented an analysis of the replies to a questionnaire on open-hearth bottom construction and materials. R. B. Snow discussed data obtained from density tests on material removed when renewing hearths. A very comprehensive paper on "Nozzle and Stopper-Head Properties, Pouring Techniques, and Their Effect on Casting-Pit Performance" was delivered by L. G. Ekholm and L. D. Hower, jun. After short addresses by J. A. Shea on "Methods of Testing Stoppers" and L. R. Berner on "Lining and Scheduling Ladles for Maximum Life," the Pittsburgh meeting closed with a discussion on the storage of refractory bricks.

The meeting at Philadelphia dealt with the maintenance and repair of basic open-hearth furnaces. C. D. Moore began by explaining how the Worth Steel Company planned their furnace-repair programme. The maintenance of roofs was discussed, and two papers were then presented, one by O. P. Luetscher on "Life and Maintenance of Open-Hearth Roofs," and the other by E. N. Hower on "Hot Patching of Open-Hearth Furnaces." Open discussions on the maintenance of bottoms, tap-hole repairs, and fettling time followed; at the end of this section of the Proceedings a summary is given of the answers to a questionnaire on the times taken for open-hearth bottom maintenance.

The Chicago meeting discussed the operation of open-hearth furnaces. The following papers were presented: "Residual Metals in Open-Hearth Steel"

by J. D. Sullivan and A. E. Pavlish; "Segregation of Scrap at Wisconsin Steel Works" by E. H. Schwartz; "A Completely Automatic Control of Open-Hearth Reversal" by B. M. Larsen and W. E. Shenk; "Stoppers and Nozzles" by J. P. L. McMahon; "Nozzle Size *vs.* Quality of Hot-Top Ingot" by H. J. Forsyth; "Factors in Mold Design for Big-End-Up Molds" by G. Soler; "Effect of Variables on Surface and Yield of Medium Carbon Semikilled Steel" by E. C. Sorrells; "Effect of Furnace, Pit, and Mold Practice on Surface and Yield of Semikilled Ingots" by J. R. Britt; "Factors Affecting Efficiency of Aluminium Additions to Fine-Grained Steel" by R. C. Diehl; "Measurement and Control of Open-Hearth Flame Radiation Intensity at Gary Steel Works" by A. L. Hodge; and "A Method of Determining Bath Temperatures" by L. F. Weitzenkorn. In addition to the above there were open discussions on improvements in charging practice, furnace-control systems, aluminium recovery, burners for liquid and mixed fuels, colloidal fuel, and flame radiation. At this meeting there was also a session on metallurgy and quality; it began with the presentation of summaries of the following papers which were to have been read at the Deoxidation Symposium in New York in February, 1945: "Effect of Deoxidation on the Strain-Sensitivity of Low-Carbon Steels" by H. K. Work and G. H. Enzian; "The Relation among Aluminium, Sulphur and Grain Size" by C. E. Sims; and "Occurrence of Oxygen in Liquid Open-Hearth Steel—Sampling Methods" by T. E. Brower and B. M. Larsen. It was continued with two papers on the economic aspects of deoxidation, one by M. Tenenbaum and the other by R. C. Diehl and F. G. Norris, and a paper by K. L. Fetters on the deoxidation of coarse-grained steels. The other papers contributed at this session were: "Elimination of Surface Defects in Low-Carbon Low-Sulphur Silicon-Killed Steel of S.A.E. 1020 Specification" by O. Pearson; "Improving Cleanness of Killed Steel" by S. Feigenbau; "Variables in Producing Clean Steel in Large Tonnages" by C. L. Labeka; "The Chemistry of Slag Control" by H. B. Emerick; "Review of Recent Developments in Basic Open-Hearth Slag Control" by M. Tenenbaum; and "Hardenability Bands and the Open-Hearth Factors Controlling Hardenability" by C. G. Atchinson. Open discussions also took place on the cleanness of killed steel, slag control, subsurface porosity, and hardenability.

The Acid Open-Hearth Technical Session was held on May 18, 1945, at Pittsburgh. One paper was presented; this was by C. E. Wenninger and W. McCaulley on the steps taken at a steel foundry to improve acid open-hearth practice. The remainder of this session was taken up with open discussions on a variety of subjects including the effect of mould life on defects on machined forgings, the sand requirements for repairing bottoms and banks, whether it is better to make additions on a rising or a falling temperature, whether additions of alloying elements should be made to the furnace or the ladle, the life of the front and back walls in 25-, 50- and 75-ton furnaces, the castability of acid open-hearth steel, the relation of furnace practice to the prevention of hot-tear formation, adding graphite instead of pig-iron to provide the carbon in acid open-hearth steel, reboiling acid open-hearth steels for forgings, the values of raw and burnt-lime additions to acid open-hearth slags.

The Manufacture of High Speed Steel. W. H. Wills. (Steel Processing, 1945, vol. 31, Nov., pp. 703–708). High-speed steel may be divided into two groups, namely, the tungsten and the molybdenum types. Electric-arc furnaces of from 3 to 6 tons are used in the melting of tool steels, the charge being melted in 2 hr. under a reducing slag. The subsequent stages of manufacture include hammer-cogging, grinding, hot-rolling, forging, annealing, and cold-drawing to sizes less than 1 in. in dia. After drawing, a final annealing is necessary before straightening; small sizes are sheared, larger bars are sawn to length.

Final Additions to an Acid Electric Heat. C. C. Wissmann. (Metal Progress, 1945, vol. 48, Dec., pp. 1296–1299). A practice for working an acid electric-furnace heat of steel is outlined with special stress on the best time to make the additions of ferro-alloys.

Industrial Electrodes. H. Sanders. (British Steelmaker, 1945, vol. 11, June, pp. 255–259; July, pp. 303–307; Aug., pp. 359–363; Sept., pp. 409–412). A comprehensive account is given of the development of the manufacture of carbon and graphite electrodes, their properties, and their application in metallurgical furnaces. Recommendations to prevent wastage in electrode consumption are made.

Big Crucible Ingots. (British Steelmaker, 1945, vol. 11, Dec., pp. 554–556). A description is given of the scene in a Sheffield melting shop in the eighteen-eighties during the casting of a 30-ton ingot from nearly 1000 crucibles each of which had been charged with 70 lb. of bar iron.

Crucible Failure in the Induction Melting Process. A. Gemant and J. Sticher. (Journal of Applied Physics, 1945, vol. 16, Nov., pp. 661–667). Experience shows that, in the course of the induction melting process, local overheating of the crucible sometimes occurs, which in extreme cases leads to failure. This is probably due to thermal instability. Overheating is more probable when the crucible material has a steep electrical-resistivity/temperature characteristic—a condition which may be caused by impurities. Computations are made showing the transition from stable to unstable conditions in the operation of an experimental and an industrial furnace and equations are developed which facilitate the selection of a current frequency which will ensure that the crucible temperature does not rise above the stable range.

Acid Open-Hearth Steel. J. R. Rait. (British Iron and Steel Research Association: Iron and Steel, 1945, vol. 18, Dec., pp. 647–654). Factors influencing the formation of non-metallic inclusions in acid open-hearth steel are discussed. The following conclusions are arrived at respecting the deoxidation of this steel: (1) The addition of silicon before manganese must be avoided because, during solution, the highly concentrated silicon streaks react with oxygen to form

refractory particles of almost pure silica; a subsequent addition of manganese is almost ineffective for deoxidation. (2) If silicon is added after manganese, it is important that the manganese/silicon ratio should not fall below 4 : 1; the time factor between the additions is therefore extremely critical. (3) The use of silicomanganese alloys, with the elements present in the correct ratio, is the surest method of producing fluid FeO–MnO–SiO_2 deoxidation products. The relative merits of finishing with a high or a low FeO content in the slag are discussed. In conclusion, casting-pit refractories as a source of non-metallic inclusions are considered.

Acid Open Hearth Slag Fluidity and Its Significance. G. R. Fitterer, J. W. Linhart, B. B. Rosenbaum, J. B. Kopec, S. Poch, and W. G. Wilson. (Acid Open Hearth Research Association, Sept., 1945, Bulletin No. 1). The Acid Open Hearth Research Association was established in 1942 by a group of American companies manufacturing acid open-hearth steel, for studying their problems by co-operative research. In this bulletin, practical methods of measuring slag fluidity and a standard fluidity test for acid open-hearth slags are described. The significance of the test is discussed. As the slag fluidity decreases during a heat, the silica content of the slag increases, the ferrous-oxide content and density decrease, and the temperature of the metal increases. A simple control procedure based on these relationships and the slag-fluidity test is described. There are two appendices, one on the theories of viscosity, and the other on laboratory determinations of slag density. A bibliography of 81 references is added.

The Reduction of Silica by Liquid Steel. W. Geller. (Iron and Steel Institute, 1946, Translation Series, No. 266). This is an English translation of a paper which appeared in Archiv für das Eisenhüttenwesen, 1942, vol. 15, May, pp. 479–490. (*See* Journ. I. and S.I., 1945, No. II., p. 184 A.)

The Origin of Silicate Inclusions in Basic Arc Furnace Steel with High Carbon Contents. A. Hultgren. (Jernkontorets Annaler, 1945, vol. 129, No. 11, pp. 633–671). (In Swedish). The occurrence of silicate inclusions in two types of steel made in a basic-lined furnace was investigated; one steel contained 1·10% of carbon and the other was a spring steel containing 0·65% of carbon, 0·70% of silicon and 1·0% of manganese. The object was to determine the conditions promoting the formation of silicate inclusions during tapping and pouring, the effect of the time the steel was kept in the ladle, and the value of certain measures for diminishing the inclusions. With the metallurgical practice followed, the steel was practically free from inclusions and low in oxygen when tapped from the furnace, but in the ladle, immediately after tapping, it contained spherical drops of silicate in greater quantity the more turbulent the tapping stream, the proportion present varying between $1{\cdot}4 \times 10^{-3}$ and 41×10^{-3} volume-%. The oxygen content also rose during tapping; contrary to expectation, the increase was less in the case of a turbulent stream. It is suggested that the intimate mixture of the white furnace slag may have caused some deoxidation of the steel; this is known as the "Perrin effect." Keeping the steel in the ladle for a few minutes caused the greater part of the silicates to disappear. For instance, the quantity was reduced from 12×10^{-3} to $1{\cdot}2 \times 10^{-3}$ volume-% in 4 min. In the same period the average diameter of the inclusions greater than 0·010 mm. decreased from 0·031–0·108 to 0·022–0·041 mm., and their structure also altered; when just formed they were mainly crystalline, whilst after a few minutes they were more or less glassy. It may be assumed that the oxidation of the tapping stream results in the formation of a silicate relatively rich in iron and manganese oxides. This will gradually dissolve in the unsaturated liquid steel and, by reaction with the silicon in the steel, the latter will become poorer in these oxides and richer in silica. At the same time the greater part of the larger inclusions will disappear from the steel by rising to the surface. Some 2-in. ingots were cast from the ladle using normal and restricted streams; the restriction caused an enormous increase in the inclusions. Evidence of the reactions between the inclusions and the steel was retained by the rapid freezing of the steel immediately after the formation of the inclusions which appeared as swarms of small drops around the larger ones. With ordinary 9 to 14-in. square ingots cast from the ladle, the silicate inclusions amounted to $0{\cdot}1 \times 10^{-3}$ to $0{\cdot}9 \times 10^{-3}$ volume-% (generally less than $0{\cdot}5 \times 10^{-3}$). Restricting the pouring stream or blowing air on to it had no appreciable effect on the quantity of inclusions, presumably because the silicate drops which did form had time to escape by rising. The last ingot to be poured was liable to contain a large number of silicate inclusions. When a Caspersson ladle was attached to the furnace and the steel was protected by an atmosphere of purified nitrogen, the silicate inclusions in the ingots were as low as $0{\cdot}02 \times 10^{-3}$ volume-%. The growth of the inclusions before the steel freezes in the mould is thought to be due to precipitation from solution and to diffusion in the liquid steel, and not to the coalescence of small drops.

FORGING, STAMPING, AND DRAWING

(Continued from pp. 22 A–23 A)

Electric-Pneumatic Hammers and Steam or Compressed Air Hammers, with Particular Reference to Their Economy. F. Knorr. (Stahl und Eisen, 1943, vol. 63, June 3, pp. 433–437; June 10, pp. 456–462). Modern electric-pneumatic hammers and steam hammers are described. The factors affecting their efficiency and data on the energy consumption of different hammers on similar production operations are presented and discussed.

Forging Die Design. J. Mueller. (Steel Processing, 1945, vol. 31, Nov., pp. 696–698). Some of the principles applied in the design of cushioned helve-

hammers and multiple dies for forging shafting are discussed.

Processing and Fabrication of Stainless Steel Sheet and Plate Products. H. S. Schaufus and W. H. Braun. (Steel Processing, 1945, vol. 31, Oct., pp. 625–629; Nov., pp. 691–694). In the first of this series of articles the composition, structure and properties of the stainless steels are dealt with, and in the second part the annealing, stabilizing and stress-relieving are discussed with micrographs of the structures obtained.

Drawing Stainless Cylinders. L. E. Browne. (Steel, 1945, vol. 117, Dec. 10, pp. 116, 117, 170). Inner cylinders which serve as vacuum chambers for water-cooled ignition tubes of stainless steel have been successfully drawn with the two ends of greater diameter than the centre, the final dimensions being accurate within 0·010 in. Details of the drawing process are given.

Cold Metal Contour Forming. R. A. MacKenzie. (Steel, 1945, vol. 117, Dec. 10, pp. 112–113, 164–166). Machines for stretch-forming and contour-forming sheet metal into a wide variety of complicated shapes are described. These machines do not require such expensive dies as do hydraulic presses.

Elastic Properties of Steel Wire Intended for Springs and Formed Parts. N. C. Talmadge. (Wire and Wire Products, 1945, vol. 20, Nov., pp. 859–861, 884). It was found that a cold-drawn wire was suitable for coiling into springs when the yield point was between 56% and 75% of the tensile strength; with higher yield-point ratios the wire was unsuitable. The effect of straightening and tempering at different temperatures on the tensile strength, yield point, and elongation of hard-drawn steel wires 0·092, 0·100, and 0·120 in. in dia. is discussed and test data are presented.

Reactive Wire Drawing. II. Janet M. Howden and R. W. Lunt. (Wire Industry, 1945, vol. 12, Dec., pp. 641–645). The available data for reactive wire-drawing are examined to derive values for the co-efficient of friction between the wire and the die, the mean pressure between the wire and the die, and a factor which is likely to determine the increase in die life which is obtained by replacing non-reactive by reactive drawing.

A Chronology of Wire and Wire Products. F. R. Morral. (Wire and Wire Products, 1945, vol. 20, Nov., pp. 862–866, 885–887). A very comprehensive chronology of the history of the development of wire making from the discovery of copper up to 1942 is presented. The chronology is the result of research work using French, German, English and American publications, and it is claimed that a complete search of all books, patents, and journals in different languages was made.

HEAT-TREATMENT

(Continued from pp. 24 A–26 A)

Neutral Salt Bath Hardening. L. F. Train. (American Society for Metals: Canadian Metals and Metallurgical Industries, 1945, vol. 8, Nov., pp. 31, 48). A description is given of a set of four electrically heated salt-bath furnaces for heat-treating tools. The four furnaces are for preheating to 1200–1750° F., bringing up to the hardening temperature in the 1750–2350° F. range, quenching in molten salt at 1200–1250° F., and tempering at 400–1100° F., respectively.

Manufacture of File Hard Precision Gears. W. G. Guthrie. (Metal Progress, 1945, vol. 48, Dec., pp. 1300–1305). A description is given of one of the General Motors Corporation plants for the heat-treatment of gears of 3·5%-nickel, 1·5%-chromium steel. The carburizing is carried out in "Homo-Carb" atmosphere-controlled furnaces.

Modern Heat Treatment of Steel. A. C. Macdonald. (Journal of the Institution of Production Engineers, 1946, vol. 25, Jan., pp. 1–28). A comprehensive account of the theory of heat-treatment is presented. Definitions of terms are given, the effects of alloying elements are explained, and descriptions are given of the Jominy end-quench test, refrigeration treatment, and induction hardening.

Modern Heat Treatment of Ferrous Materials. J. R. Mott. (American Society for Metals: Canadian Metals and Metallurgical Industries, 1945, vol. 8, Nov., pp. 26–28). The preparation of atmospheres for carburizing steel is discussed and methods of determining whether the surface of steel has been carburized or decarburized are described.

Surface Hardening. Possibilities of its Application to Castings. M. B. de Jessey. (Iron and Steel, 1945, vol. 18, Dec., pp. 667–669). This is a translation of an article published in "Bulletin de l'Association Technique de Fonderie," dealing with the surface hardening of iron castings. The process consists of raising a part of the metal locally to the hardening temperature and quenching it, the structure and constitution of the rest of the casting remaining unchanged. Two methods of heating are in use: *viz.* by induction and by blowpipe. The hardening technique is fully described.

The Development of Induction Furnaces and Physical Laws. L. Dreyfus. (Teknisk Tidskrift, 1946, vol. 76, Feb. 2, pp. 113–123). (In Swedish). The theory governing the design of high-frequency induction furnaces is critically discussed with special reference to furnaces for hardening and tempering rock-boring bits and gear teeth, examples of which are described.

Induction Heating. J. S. Edgar. (American Society for Metals: Canadian Metals and Metallurgical Industries, 1945, vol. 8, Nov., pp. 28–30). Several examples of the application of induction hardening are described.

High-Frequency Heating. (Automobile Engineer, 1946, vol. 36, Jan., pp. 13–19). The basic principles of induction heating, the fields of application, and efficiencies are outlined. The three types of equipment, motor-generator, spark-gap converter, and thermionic oscillator are discussed. Special-purpose machines for through-heating, surface hardening,

semi-continuous operation, and localized heating are described in detail.

Effects of Various Surface and Structural Conditions on Nitriding. R. W. Allott. (Machine Shop Magazine, 1945, vol. 6, Dec., pp. 90–92). The author disagrees with the conclusions drawn by F. F. Dodson (*see* Journ. I. and S.I., 1945, No. II., p. 139 A) relating to the causes of defects in nitrided cases. Experiments are described which support the view that defects in nitrided cases are not necessarily due to the presence of grease or oxide films.

Some Aspects of the Overheating of Steel Drop-Forgings. H. J. Merchant. (Iron and Steel Institute, 1946, this Journal, Section I.). The effects of overheating on the serviceability of steel drop-forgings, particularly in relation to their use in aero-engine construction, are discussed. The principal causes of, and factors influencing, overheating are described, and an attempt is made to differentiate between overheated steel, severely overheated steel, and burnt steel. The greater part of the paper is devoted to a description of the methods used to detect overheating in alloy steel, namely, examination of the surface and of the fracture, metallographic examination, and mechanical testing. Work carried out with a view to studying obscure aspects of the subject, together with an attempt to establish fracture-test standards, is described. An account is given of methods of preventing overheating and of reclaiming overheated steel. Finally, there are some notes on the occurrence of overheating in American steels, and some suggestions are given regarding directions which might be taken by future research on this subject.

The Overheating and Burning of Steel. A. Preece, A. Hartley, S. E. Mayer, and J. Nutting. (Iron and Steel Institute, 1946, this Journal, Section I.). The overheating properties of a wide variety of steels have been examined. The temperatures of overheating have been determined on the basis of the fracture and etch tests, a critical examination of these tests showing them to be satisfactory for this purpose.

There would appear to be no clear relationship between the overheating temperature and the normal specified analysis of the steel, or its oxygen and nitrogen contents. Open-hearth steels have generally a higher overheating temperature than electric steels. An unexpected correlation has been observed between the inclusion content of a steel and its overheating temperature.

Furnace conditions, such as composition of the atmosphere, soaking time, and the rate of cooling, have been studied in relation to overheating. From this, the possibility of a slow cooling treatment as an effective method of restoring overheated steels has been indicated. The effect of overheating upon the tensile and Izod properties has also been examined.

Results obtained in this investigation provide further information on the mechanism of overheating.

The Effect of Oxygen on the Isothermal Transformations of Steel, and a Suggested Test for Burning. F. C. Thompson and L. R. Stanton. (Iron and Steel Institute, 1946, this Journal, Section I.). Samples of a series of plain carbon steel have been locally burned, and subsequently isothermally quenched to various temperatures, with a view to finding a test for burned material.

The results show that certain steels are rendered more reactive by burning, and this suggests a displacement of the inherent S-curve; the nature of the final product of transformation indicates that this displacement is in a vertical direction, which in effect acts as a shift to the left.

The deduction is made that this behaviour is due to the introduction of oxygen during burning, which probably acts indirectly.

The effect is absent in other steels of similar composition, as determined by ordinary methods of analysis, and in general it is found to be too erratic to offer a universal test for burning.

Overheating and Burning of Nickel-Chromium-Molybdenum Steel. W. E. Goodrich. (Iron and Steel Institute, 1946, this Journal, Section I.). Occasionally, fractured test-pieces removed from forgings for routine inspection have disclosed small areas of granular facets, which some inspectors have interpreted as indicative of "burnt" steel. More conservative opinion, however, has regarded these areas as possible signs of "overheating," and that the evidence of such a condition might be masked by slow rates of cooling while, on the other hand, signs of overheating, as judged by fractures, may occur in material which has not, in fact, been overheated. A series of experiments were therefore conducted on sections machined from a nickel–chromium–molybdenum–vanadium steel forging, the sections being heated to initial temperatures varying between 1250° and 1400° C. and, afterwards, subjected to various cooling treatments, in some cases followed by a further standard oil-hardening and tempering operation. Afterwards, the specimens were fractured and their fractures and metallographic structures, in the various heat-treated conditions, are described.

It has been shown that a repetition of normal oil-hardening and tempering treatments, or slow cooling from overheating temperatures, can appreciably reduce, and in many cases entirely eliminate, the abnormal granular type of fracture. All the evidence obtained indicated that it was necessary to cool fairly rapidly through the upper critical range to create conditions favourable for the formation of the granular facets, and even then tempering temperatures exceeding approximately 400° C. were necessary to produce the relative strengths of the crystals and crystal boundaries required for their formation.

Even at overheating temperatures of the order of 1400° C., no incipient fusion was detected at the grain boundaries or within the grains, though there were signs of a metallographic change apparently taking place in the solid state at the grain boundaries and also originating from non-metallic nuclei within the grains.

Some Experiments on Overheating. J. Woolman and H. W. Kirkby. (Iron and Steel Institute, 1946, this Journal, Section I.). A series of experiments has been carried out which were designed to throw light on the phenomenon of overheating and the

factors affecting it. Some evidence is submitted in support of a hypothesis that the overheating phenomenon indicated by facet fractures is a precipitation effect and is not the result of external gas atmospheres. It is considered that the precipitate is already present in the steel as cast, and the effect of heating to a high temperature is to redissolve it. Subsequent cooling, if at a suitable rate, results in a precipitate forming at the existing austenite boundaries which then persist as a network. This is reflected in the final fracture test by interruptions in the final path of the crack, giving rise to the facet type of fracture. The indications are that the rate of cooling from the overheating temperature is an important factor in producing facet fractures. The method of steelmaking and the addition of aluminium and other killing agents also have an influence on the minimum overheating temperature, although the reasons for both effects are not yet clear. The vacuum-melting experiments have also shown that a material having a very high overheating temperature can be substantially reduced by remelting, although the variables due to changes in composition introduce complexities. Besides oxygen, the sulphur contents appear to play an important rôle in the phenomenon.

WELDING AND CUTTING

(Continued from pp. 26 A–27 A)

An Analytical Examination of Weld Deposits from Commercial Mild Steel Electrodes to Specification B.S.S. 639A, and the Manufacture of Artificial Weld Metal. G. L. Hopkin. (British Welding Research Association, Symposium on Metallurgy of Steel Welding, Nov., 1945). A series of deposits from twenty commercial mild steel electrodes have been examined analytically. Large variations were found in the nitrogen contents of samples taken from top and bottom of deposits made with 12 S.W.G. electrodes, but the differences decreased as the electrode size increased and in 50% of the 4 S.W.G. electrodes the bottom of the deposit showed a higher content than the top. Large variations in silicon content were found between the different gauges, the 12 S.W.G. electrodes invariably containing the highest content. To study the effect of nitrogen on the mechanical properties of cast steel of weld-metal composition, attempts were made to manufacture an artificial weld metal high in nitrogen. It was found, however, that exceedingly long annealing times would be required in order to obtain diffusion of the nitrogen to the centre of the specimens; this investigation was therefore discontinued.

Comments on the Rôle of Hydrogen in Relation to the Cracking of Alloy Steels on Welding. G. L. Hopkin. (British Welding Research Association, Symposium on Metallurgy of Steel Welding, Nov., 1945). The theories put forward to explain why the heat-affected zones of parent plate vary in their apparent brittleness when different welding electrodes are used are discussed. The main factor, according to the author's theory, is the quantity of hydrogen which diffuses into the somewhat brittle and stressed heat-affected zones during cooling and when cold. Tests have shown that the killed ferritic deposits evolve two to three times the volume of hydrogen in the cold which is evolved by specially made low-hydrogen electrodes, and that these, in turn, evolve about two to three times the hydrogen given off from the "oxidised" deposits. Little or no gas is evolved from the austenitic deposits.

The Effect of Initial Heating Temperature on the Mechanical Properties of an Air-Hardened Ni–Cr–Mo Steel. J. A. Wheeler and V. Kondic. (British Welding Research Association, Symposium on Metallurgy of Steel Welding, Nov., 1945). A report is presented of an investigation of the effect of the temperature to which an air-hardening steel is heated before quenching, on the mechanical properties of the martensite which is formed. The temperature range covered was 850–1475° C., which includes almost the whole range of heat-treatment temperatures characteristic of the hardened zone of a weld. The steel used contained 3·38% of nickel, 0·65% of chromium, and 0·26% of molybdenum. There was evidence of a slight abnormality in the region of 1350° C. as there was a drop in the hardness and tensile strength after initial treatment at this temperature.

Electro-Physics of the Welding Arc. L. H. Orton. (British Welding Research Association, Symposium on Metallurgy of Welding, Nov., 1945). Work already done on the physics of the welding arc is reviewed and the programme of work to be carried out by the Committee on "The Electro-Physics of the Arc" is outlined.

Cracking of Welded Gas Mains. L. Reeve. (British Welding Research Association, Symposium on Metallurgy of Welding, Nov., 1945). The causes of failure of welded gas mains by cracks appearing near butt and fillet welds are being investigated. Similar cracks have been developed in welded steel plate immersed in boiling solutions of ammonium, calcium, and sodium nitrate in as short a period as 16 hr., and the following tentative conclusions have been arrived at: (1) A minimum degree of stress, which varies with the type of steel used, must be present to produce failure; the most sensitive steels will fail at an imposed stress of less than 4 tons per sq. in. in the absence of plastic strain. (2) With severe plastic strain, failure will occur within 10 days with all the types of steel that have been tested. (3) Failure can be prevented by stress-relieving treatment.

An Apparatus for the Determination of the Viscosity of Welding Slags. P. K. Gledhill. (British Welding Research Association, Symposium on Metallurgy of Steel Welding, Nov., 1945). A description is given of a modified Herty-type viscometer with which rapid determinations of the viscosity of welding slags can be made.

Statement on Weldability Tests for Specification Purposes. H. O'Neill. (British Welding Research Association, Symposium on Metallurgy of Steel Welding, Nov., 1945). A brief survey is made of the work done by Committee FM.2 of the British Welding Research Association which is studying weldability

tests. All known tests have been reviewed, but only the following are receiving further consideration: (*a*) The tube triangle test; (*b*) the circular patch test; (*c*) the Reeve fillet test; and (*d*) the rigid butt joint test.

The Flash Welding of Steel. J. C. Barrett. (Iron and Steel Engineer, 1945, vol. 22, Dec., pp. 103–113). The principles of flash-welding are discussed and their applications to modern production problems are described. Many illustrations of machines for flash-butt-welding bars up to 1¾ in. in dia., aircraft landing gear assemblies, and crankshaft counterweights are reproduced.

Production Technique and Quality of Flash-Welded Joints. H. Kilger. (Welding Journal, 1945, vol. 24, Aug., pp. 413-S–432-S; Oct., pp. 506-S–520-S). An English translation is presented of sections of Part II. of Dr. Hans Kilger's book "*Fertigungstechnik und Güte abbrenngeschweisster Verbindungen*" which was published in 1936. Although the work is limited to manually operated machines and to the simpler commercial carbon steels, the author has considered, and to some extent solved, many of the problems confronting the welding industry.

Observations on the Appearance Welding of Malleable Castings. H. A. Schwartz, I. Young, and J. Hedberg. (American Society for Testing Materials, 1945, Preprint No. 31). The repair of malleable iron castings by welding is discussed with special reference to the microstructure which is to be expected after employing different welding techniques. It is considered that, with a short heat-treatment after welding, a machinable repair can be effected.

The Distortion of Steel Joists Welded on One Flange. R. Malisius. (Zeitschrift des Vereines deutscher Ingenieure, 1940, vol. 84, Sept. 7, pp. 661–662). Accurate measurements were made of the distortion of flat steel bars and joists caused by the deposition of weld metal along one edge or flange. The effects of different thickness and width of the parent metal on the hogging and sagging during and after welding was studied.

Straining of Weld Metal during Cooling. E. P. De Garmo. (Iron Age, 1945, vol. 156, Dec. 20, pp. 48–50). Failures have occurred in welded joints where the weld metal is deposited on plates offering a high degree of restraint, and in these cases the weld metal exhibited no ductility. Specimens were cut from different positions along a butt weld 2 ft. long joining two large plates 1 in. thick. Those from the ends of the weld, which had not been plastically strained during cooling, were given known amounts of plastic strain. Tensile tests were then made on all the specimens and the yield points, maximum and ultimate loads and total elongations were very accurately determined. A comparison of the results revealed that the metal half-way along the weld had been strained plastically by an amount equivalent to a stress higher than the yield point.

Effect of Three Variables in Spot-Welding. H. Van Sciver. (Metal Progress, 1945, vol. 48, Dec., pp. 1292–1295). The strength and structure of spot welds in 18/8 stainless steel sheet 0·075 in. thick was investigated by series of tests in which three of the four conditions, viz., current, time, tip pressure, and electrode shape, were kept constant and the fourth was varied. The results are indicated by a series of 32 photographs of sections through the welds.

Weldability. A. Leroy. (Revue de Métallurgie, Mémoires, 1944, vol. 41, Jan., pp. 26–32). A short review is presented of welding methods and tests, and factors influencing the weldability of metals are discussed.

Weldability Factors for 18-8 Stainless. F. H. Page, jun. (Iron Age, 1945, vol. 156, Dec. 27, pp. 58–61). Different heats of stabilized 18/8 stainless steel are known to show wide differences in weldability even when their chemical composition complies with official specifications. The effects of slight variations in the percentages of silicon, manganese, and other elements on the weldability are discussed.

The Fatigue Strength and Notch Sensitivity of Electric-Arc Weld Metal. C. Schaub. (Teknisk Tidskrift, 1945, vol. 75, Nov. 17, pp. 1263–1266). (In Swedish). The results of rotating bend-fatigue tests on specimens of weld metal machined with different radii at the shoulder are presented and discussed. These show that the notch sensitivity of weld metal is greater than is generally believed.

The Production and Repair of High-Speed Steel Tools by Welding. R. Gunnert. (Teknisk Tidskrift, 1946, vol. 76, Jan. 19, pp. 57–64). (In Swedish). A comprehensive account is given of the production, heat-treatment, and repair of high-speed steel tools and milling cutters by the deposition of tungsten-steel weld metal. The influence of different heat-treatments on tool life is discussed.

The Cold-Welding of Cast Iron. P. Krug. (Zeitschrift des Vereines deutscher Ingenieure, 1940, vol. 84, Oct. 12, pp. 777–783). Hardness and tensile tests were carried out on electric-welded joints in five different cast irons. The joints were cold-welded (*i.e.*, without preheating) and the effect of spraying the joint faces with aluminium, lead, zinc, nickel, and copper was studied. Coating with copper greatly improved the quality of the welds obtained in irons with poor welding properties. More than 1% of phosphorus in the cast iron has a detrimental effect on the weldability.

Repair of Rolling-Mill Pinion by Thermit Welding. (Engineering, 1946, vol. 161, Jan. 11, pp. 32–33). An account is given of the technique employed for the repair by thermit welding of a broken pinion shaft for a rolling mill. The broken portion of the 27-in. dia. shaft was parted off and a new length of shaft 27¼ in. in dia. was successfully welded on by the thermit process.

Aluminium Bronze Welding in Steel Plant Maintenance. F. E. Garriott. (Iron and Steel Engineer, 1945, vol. 22, Dec., pp. 98–102). The application of aluminium-bronzes (*i.e.*, copper alloys containing 5–14% of aluminium and up to 5% of iron) in steel mills for repairing equipment subject to wear and corrosive attack is discussed and the mechanical properties of the alloys detailed. Their application in the form of coated electrodes for metallic-arc, carbon-arc, and oxy-acetylene welding to make joints of high strength for maintenance work is dealt with.

Production Brazing Developments. (Machinery, 1946, vol. 68, Jan. 17, pp. 77–84). After a historical review of brazing and soldering methods the modern technique is described and illustrated, and the post-war possibilities in production discussed. Various applications are described, among them the fabrication and repair of tools. Whilst the older methods required skilled operators, the modern high-production methods tend to eliminate the human element by employing automatic or semi-automatic means of heating (*e.g.*, by induction). The silver-copper-cadmium-zinc alloy is mainly used in the form of wire, rod, or sheet in a wide variety of sizes. The importance of cleaning is emphasized.

New "Heenan" Plant; Chain Making Equipment. (Wire Industry, 1945, vol. 12, Dec., pp. 637–639). Modern machinery for forming and welding chains is described.

Powdered Metal Welding Rods. (Steel, 1945, vol. 117, Dec. 10, pp. 130, 133). A new process is described, which was developed for making electrodes for welding alloy steels by compacting powdered metals around an iron wire core.

Standardization in Welding. (Zeitschrift des Vereines deutscher Ingenieure, 1940, vol. 84, May 11, pp. 325–329). Particulars are given of the proposed German standard specifications for welding processes. Draughtsmen's symbols for welded joints, which are intended to bring up-to-date the DIN 1910/12 specifications issued in 1927, are also given.

Oxy-Acetylene Cutting. P. Krug. (Zeitschrift des Vereines deutscher Ingenieure, 1940, vol. 84, Sept. 21, pp. 713–716). Some German oxy-acetylene cutting machines are described and data on cutting speeds, consumption of oxygen and acetylene, and the effect of alloying elements on the cutting process are presented and discussed.

MACHINING

(Continued from p. 27 A)

Machinability. (Automobile Engineer, 1945, vol. 35, Dec., p. 553). A discussion of machinability, defined as the property of the material that affects the speed at which a given cut may be taken, and its importance in production economics, is presented. It is shown by machining experiments that the structure of the metal has a marked influence on the machinability of steel. The degree of machinability (in turning, forming, drilling, and broaching) of steels heat-treated in various ways is given in a table.

Machinability of Sulphurized Steels. H. M. Clarke. (Steel, 1946, vol. 117, Dec. 17, pp. 116–119, 162). Various processes of adding sulphur to steel to give it good machining properties have been patented. In this paper an addition in the form of an anhydrous sulphite is advocated. The results of machining tests on steels sulphurized by this method are presented and discussed.

The Measurement of Finely Finished Surfaces by Optical Interference. C. Timms. (Journal of Scientific Instruments, 1945, vol. 22, Dec., pp. 245–246). The author describes a method of measuring the depth and pitch of fine surface irregularities as produced by lapping or similar refined processes, by making use of optical interference. It is strictly limited to fine finishes where the surface irregularities do not exceed 0·00001 in. total depth. The method is self-calibrating since the separation of any two adjacent fringes automatically provides a scale of units each corresponding to half the wave-length of the light used.

Optical Methods of Examining the Quality of Surfaces. B. Frischmuth. (Schweizer Archiv, 1945, vol. 11, Sept., 262–269). Optical methods of evaluating metal surfaces are reviewed and a description is given of an instrument which, in conjunction with a metallographic microscope, enables evaluations to be made by the light-interference method.

CLEANING AND PICKLING OF METALS

(Continued from pp. 9 A–10 A)

Glycerine in Metal Treatment. G. Leffingwell and M. A. Lesser. (Engineering Materials and Processing Methods, 1945, vol. 22, Nov., pp. 1432–1434). The literature on the use of glycerine in solutions for cleaning metals is reviewed.

The Use of Steel Turnings as a Cleaning Medium Instead of Steel Grit. A. Kentischer. (Giesserei, 1944, vol. 31, July, pp. 101–106). Owing to the difficulty in obtaining steel grit for cleaning castings tests were made using cast-iron borings, chips from the planing of cast iron, and steel turnings. All three materials were found to be good substitutes. When steel turnings were broken up by tumbling with small castings in a revolving drum, the product had about the same particle size as the steel grit and it had a longer life when used for cleaning by the centrifugal method. A great saving was effected, as the broken steel turnings cost only 15% of the price of the steel grit.

Cleaning Castings by the Hydro-Blast System. (Machinery, 1945, vol. 67, Dec. 13, pp. 672–674). A detailed description is given of the equipment for cleaning castings by the Hydro-blast system in which a stream of water and sand at high pressure (1200 lb./sq. in.) is projected against the casting from a gun.

Electrolytic Methods of Polishing Metals. S. Wernick. (Sheet Metal Industries, 1945, vol. 22, July, pp. 1221–1222, 1229–1233; Sept., pp. 1586–1592; Nov., pp. 1951–1958). Conclusion of a series of

articles (*see* Journ. I. and S.I., 1945, No. II., p. 61 A). In Part VIII. the polishing of the precious metals is described. Part IX. is devoted to the polishing of rhodium. In Part X. the literature on the electrolytic polishing of carbon steels is reviewed.

Selection of Metal Cleaning Methods. C. D. Townsend. (Engineering Materials and Processing Methods, 1945, vol. 22, Nov., pp. 1411–1417). Petroleum solvents, emulsifiable cleaners, alkaline cleaners, vapour-degreasing, blast-cleaning and tumbling are discussed and the characteristics of each process are pointed out.

Vapor Degreasing. J. C. Joyce. (Steel, 1945, vol. 117, Nov. 26, pp. 110–116). Descriptions are given of modern electric and gas-heated degreasing plants employing perchlorethylene or trichlorethylene.

Sodium Hydride Descaling. L. W. Townsend. (Steel, 1945, vol. 117, Nov. 12, pp. 122–123, 180–184). A problem in the production of stainless clad steel was to obtain a white pickled stainless steel surface and at the same time prevent the over-pickling of the low-alloy backing steel. The bath prepared for this purpose consisted of molten caustic soda at 700° F. into which metallic sodium and hydrogen were introduced to form a solution of 2–3% sodium hydride. After hosing the steel with water at high pressure it is dipped for 2 min. in a 5–10% sulphuric acid bath followed by immersion for 2–3 min. in 10% nitric acid at 160° F.

Descaling Stainless Steel. (Steel, 1945, vol. 117, Nov. 19, pp. 121, 180–182). The application of the sodium-hydride descaling process (*see* preceding abstract) to the cleaning of coils of stainless steel wire is described and illustrated.

Sodium Hydride Process for Descaling Steel. J. Albin. (Iron Age, 1945, vol. 156, Nov. 8, pp. 58–63). The development of the plant and process for descaling alloy steels, and especially stainless steel, in a solution of sodium hydride and caustic soda is described (*see* preceding abstracts).

The Effects of Different Methods of Pretreating Iron and Steel before Painting. F. Fancutt. (Iron and Steel Institute, 1946, Special Report No. 31). The dependence of the durability of paint applied to iron and steel upon the surface condition is strongly emphasized by the results of this investigation. All treatments tested which leave the steel surface in a rust-free, scale-free condition at the time of paint application are comparable in presenting an "ideal" surface for painting, and the choice between them will generally be determined on economic grounds or convenience of operation.

The simultaneous presence of rust and scale below a paint film induces rapid breakdown, while light surface rust in the absence of scale, though accelerating paint failure, is not so deleterious, especially where red-lead paint is used in priming. Exposure of the descaled steel, unless accompanied by the formation of a visible film of rust, does not, however, shorten the life of paint subsequently applied.

The durability of paint applied to the as-rolled surface (*i.e.*, scale bearing, but rust-free), provided the film of scale is practically complete and unbroken, is much the same as for the scale-free, rust-free surface.

The rate of breakdown of paint applied to a clean, rust-free, scale-free surface appears to be mainly a function of the paint itself, and to be largely independent of (*a*) the type of steel, (*b*) the descaling method employed, *i.e.*, chemical or mechanical, (*c*) nature of acid used, (*d*) the presence or absence of inhibitors added to the pickling bath, and (*e*) the washing process employed after pickling.

The life of paint applied to a steel surface not in the ideal condition is, on the other hand, influenced by a number of factors, including (*a*) the type of steel, and (*b*) the nature of acid used, in addition to the sometimes overwhelming factor of surface condition.

The method of paint application is of importance when it influences the weight of paint applied, and in this investigation spray-applied films were heavier and consequently more durable than brushed ones. Red-lead priming is much superior to red oxide on rusty surfaces, but the difference is less, and may even be reversed under certain conditions, when rust-free, scale-free surfaces are involved.

Flame-Priming Steel Surfaces for Painting. R. J. Anderson. (Engineering Journal, 1945, vol. 28, Nov., pp. 707–712). The behaviour of rust and mill-scale on heating is discussed and the flame-priming (heating with a multiple-burner oxy-acetylene torch) of steel surfaces to prepare them for painting is described. A burner head 6 in. wide has 49 flame ports $\frac{1}{8}$ in. apart. If the burner is kept moving at 35 ft./min., about 13 cu. ft. each of oxygen and acetylene would be consumed for every 100 sq. ft. of surface.

A Mixed Acid Pickling and Recovery Installation Based on the De Lattre Process. E. W. Mulcahy. (Sheet Metal Industries, 1946, vol. 23, Jan., pp. 77–81). A description of the De Lattre pickling process has been given by A. McLeod (*see* Journ. I. and S.I., 1940, No. II., p. 189 A); since then two plants have been installed in Great Britain and a description of the larger one is given in this paper. Special advantages of the process are that the recovery method has reduced the acid consumption to a minimum; and the steam consumption also is low owing to the lower working temperatures required.

COATING OF METALS

(Continued from pp. 10 A–11 A)

Plating with Insoluble Anodes. W. R. Binai. (Metal Finishing, 1945, vol. 43, Apr., pp. 144–146, 147). The advantages and disadvantages of plating with insoluble anodes are discussed.

Electroplating Metal Sheets Prior to Fabrication. E. A. Ollard and E. B. Smith. (Sheet Metal Industries, 1946, vol. 23, Jan., pp. 116–126). Details are given of the production of electroplated steel strip for making pressings and stampings.

An Examination of Electroplating Processes for

Stainless Steels. H. Krause. (Korrosion und Metallschutz, 1940, vol. 16, Sept., pp. 304–308). Old and new methods of plating stainless steels are reviewed.

Hard Chromium Plating. J. J. Dale. (Monthly Review of the American Electroplaters' Society, 1945, vol. 32, June, pp. 581–594, 619). A general survey of the technique and use of hard chromium plating is presented. Operating characteristics of the baths, pretreatments for the metals to be plated, and practical advice to ensure good results are included.

The Application of Electrolytic Chromium as Protection Against Corrosion. K. Gebauer. (Korrosion und Metallschutz, 1940, vol. 16, Sept., pp. 297–299). Methods of plating chromium on iron and steel by electrolysis are discussed, with special reference to the conditions which prevent the formation of porous or cracked coatings.

Nickel Flashing and Its Relation to Enamel Adherence. E. Wainer and W. J. Baldwin. (Journal of the American Ceramic Society, 1945, vol. 28, Nov., pp. 317–326). The manner in which a flash coat of nickel improves the adherence of enamel on steel is discussed. The rôle of nickel in improving adherence is possibly bound up with retarded oxidation. Adherence may be considered as a corrosion phenomenon developed by the action of gases and other reagents. The degree of adherence is a function of the oxygen pressure at the interface when the glass is fused. The equilibria developed are strongly affected by the presence of nickel. Using coatings directly on iron, the amount of nickel required varies with each particular enamel composition. Some evidence regarding the nature of the oxide that promotes adherence is presented.

The Technique of Sheet Galvanising by the Hot Dip Process. H. Edwards. (Sheet Metal Industries, 1945, vol. 22, Oct., pp. 1725–1730, 1736; Nov., pp. 1914–1922; Dec., pp. 2096–2103, 2108). Conclusion of a series of articles on the hot-dip galvanizing process as practised at the works of John Summers and Sons, Ltd. (*See* Journ. I. and S. I., 1945, No. II., p. 166 A.

Specification Zinc Plating on Steel Sheets. W. H. Safranek. (Monthly Review of the American Electroplaters' Society, 1945, vol. 32, June, pp. 567–578). The conditions which are essential for good zinc-plating on steel sheet for assemblies requiring fairly high tensile strength without reduction in ductility and corrosion resistance, are discussed. A number of graphs and tables correlate the effects of various chemicals, temperature, current density, and dilution with efficiency and throwing power.

Current Zinc Electroplating Practice. A. G. Gray. (Steel, 1945, vol. 117, Oct. 8, pp. 109, 158–170; Oct. 15, pp. 136–142; Oct. 22, pp. 129–134; Oct. 29, pp. 110–114, 146; Nov. 5, pp. 142–144, 182–186; Nov. 12, pp. 132–136). Detailed descriptions of electrolytic methods of depositing zinc on steel are given; the merits and limitations of acid and of cyanide baths are discussed with data on these methods as well as on the zinc-mercury process. The testing and properties of the coatings are also dealt with.

The Cronak Process as Applied to Zinc Plate. S. E. Maxon. (Metal Finishing, 1945, vol. 43, Apr., pp. 148–149). A method of increasing the corrosion resistance of zinc coatings is described. It is called the Cronak process, and consists of cleaning the zinc surface and dipping in a sodium-dichromate–sulphuric-acid solution.

Developments in Tin and Tin Alloy Coatings. E. S. Hedges. (Chemistry and Industry, 1945, Dec. 1, pp. 370–372). Some results of recent work at the Tin Research Institute are reviewed. The most important single factor in obtaining high quality tin coatings is the preparation of the steel base. Provided that rigorous precautions in the preparation are taken, acid-stannous-sulphate baths produce the best quality coatings; the alkaline bath deposits are considerably less sensitive to the pretreatment. Two methods of tinning cast iron are briefly described.

The Tinning of Cast Iron. R. A. Cresswell. (Journal of the Iron and Steel Institute, 1945, No. II., pp. 157 P–169 P). The quality of hot-dipped tin coatings on cast iron after different pretreatments is investigated, using the criteria of continuity of coating, rust resistance, and adhesion. The nature of the bond between tin and cast iron is also studied by microscopical examination.

The poor tinning quality of cast iron is shown to be caused mainly by the presence of graphite flakes and to be intensified by contamination of the iron surface by graphitic sludge after pickling in acids. Investigation of modifications of the ordinary tinning processes shows that improved tin coatings can be obtained by reducing the duration of pickling in acids to a minimum; this is facilitated by good mechanical cleaning. The best pickling agent is found to be a cold mixture containing 5 vol.-% sulphuric acid and 5 vol.-% hydrofluoric acid. The corrosion resistance and adhesion of coatings obtained by such modifications of ordinary methods are, however, still low.

Two new methods of preparing cast iron for tinning are described, both involving treatment in molten salts. The chloride method consists of immersing the article in the fused zinc-chloride–sodium-chloride eutectic mixture at 300–350° C., followed by dipping in tin covered with a layer of the same fused salt mixture, at 300° C. This method gives adherent protective coatings of tin on cast iron which has been well cleaned mechanically. When the cast iron is bonded to white metal, consistent adhesion values of from 2 to 2·8 tons/sq. in. are obtained. The chloride method can be adapted to tinning by wiping.

The nitrate method was developed to give highly adherent coatings of tin and involves the following steps: light pickling in acid to "open up" graphite flakes; immersion for 15 min. at 300–350° C. in fused sodium-nitrate–potassium-nitrate eutectic mixture in order to oxidise surface graphite; pickling in dilute hydrofluoric acid to remove the scale formed in the nitrate bath; fluxing and tinning. Adhesion values of up to 3·8 tons/sq. in. are obtained between tin-base bearing alloy and cast iron tinned by this method. Microscopical examination of the bond shows that the tin penetrates cavities formerly occupied by graphite.

Electrodeposition of Iron-Tungsten Alloys from an Acid Plating Bath. M. L. Holt and R. E. Black. (Electrochemical Society: Metal Finishing, 1945, vol.

43, Apr., pp. 150–153, 176). A report is presented on the possibilities of electrodepositing tungsten alloys from commercial plating baths which are modified by an addition of sodium tungstate.

Metallizing as a Production Process. (Iron Age, 1945, vol. 156, Dec. 13, pp. 72–73). The utilization of the metallizing process in production methods is outlined and a description of the latest type metal-spraying gun is given.

Flame Spraying. W. D. Jones. (Metal Industry, 1946, vol. 68, Jan. 25, pp. 63–65). The process of protecting iron and steel surfaces by spraying metal powder with the Schori pistol is described. The advantages of this type of flame pistol are listed, and the application of zinc and aluminium for corrosion protection is outlined. One great advantage of the gun is its ability to apply a wide variety of non-metallic materials. The difference between flame-spraying and air-spraying is that in the former case one can dispense with solvents and drying, and thick coatings can be produceed in one application.

Corrosion Protection by Phosphatization—A Review. M. A. Streicher. (Corrosion, 1945, vol. 1, Dec., pp. 219–227). This review covers the chemistry of phosphatization, the properties of the solution and the phosphate coating, and industrial applications of the process. There are 43 references.

Physical Properties of Sprayed Metals. A. P. Shepard. (Metco News : Welding Journal, 1945, vol. 24, Oct., pp. 937–938). The structure of sprayed metal coatings is discussed, and data on the specific gravity of a number of sprayed coatings of ferrous and non-ferrous alloys are presented and compared with that of the wire used in each case.

Nomograph on Thicknesses of Electrodeposits. A. S. Covert. (Metal Finishing, 1945, vol. 43, Apr., p. 147). A nomograph is presented which enables a rapid determination to be made of the plating time necessary to produce a given thickness when the metal and the current density are known.

End Point Indication of the B.N.F. Jet Test for Measurement of Thickness of Zinc Coatings on Steel. S. G. Clarke. (Journal of The Electroplaters' Technical Society, 1945, vol. 20, p. 75). Particulars are given of a potassium ferricyanide solution which gives a clear indication of the perforation in the British Non-Ferrous Metals Research Association test for measuring the thickness of zinc coatings on steel.

The Surface Treatment of Hot-Dip Galvanised Coatings Preparatory to Painting and Its Relation to Corrosion Resistance. E. F. Pellowe and F. F. Pollak. (Sheet Metal Industries, 1946, vol. 23, Jan., pp. 82–85). Methods of improving the adhesion of paint coatings on galvanized steel are discussed; these include natural weathering, mechanical roughening, etching, and phosphatizing.

PROPERTIES AND TESTS

(Continued from pp. 27 A–31 A)

Developments in Ferrous Metallurgy. C. Sykes. (Institution of Civil Engineers, James Forrest Lecture : Metal Treatment, 1945, vol. 12, Winter Issue, pp. 245–258). The more outstanding developments of the last twenty years in the production of steel and in the knowledge of transformations, heat-treatment, welding, corrosion, scaling resistance and high-temperature properties are reviewed.

Metallic Materials Inspection with Particular Reference to Non-Destructive Testing Methods. J. E. Garside. (Manchester Association of Engineers, Nov., 1945, Advance Copy). A comprehensive survey is made of non-destructive methods of testing metals with special reference to those requiring neither expensive equipment nor highly-trained operators.

Non-Destructive Testing of Steel Castings. (American Foundryman, 1945, vol. 8, Nov., pp. 35–36). Methods and progress in setting up standards for the non-destructive testing of steel castings are reported. A bibliography of 20 recent publications is appended.

Contribution to a Study of the Influence of Austenite on the Mechanical, Physical and Technical Properties of Carbon Steels. G. Delbart and R. Potaszkin. (Revue de Métallurgie, Mémoires, 1944, vol. 41, Apr., pp. 97–110; May, pp. 143–149). A detailed investigation has been made of soft and medium carbon steels, prepared in the Héroult electric furnace. Data are given of the melting process and deoxidation, the latter being effected by different aluminium additions for each cast. The McQuaid-Ehn test was used for the grain-size determination. Test-pieces for the mechanical testing were taken from 30-mm. bars forged from 500-kg. ingots. The influence of the grain size of the austenite on the mechanical properties of soft carbon steel after annealing, case-hardening and quenching, and of medium-carbon steel after annealing and thermal refinement is discussed. The effect which the austenitic grain size has on the transformation points and on the ability of steel to undergo plastic deformation was also investigated.

The Effect of Combined Stresses on the Mechanical Properties of Steels between Room Temperature and −188° C. D. J. McAdam, jun., G. W. Geil, and R. W. Mebs. (American Society for Testing Materials, 1945, Preprint No. 22). Tensile tests were made on unnotched and notched specimens of carbon steels and alloy steels at several selected temperatures between room temperature and that of liquid air (−188° C.). The results are compared with those obtained with some non-ferrous metals. Results are presented in 17 diagrams which reveal the influence of combined stresses and stress concentration on the mechanical properties at various temperatures. Additional diagrams show the quantitative variation of these properties with temperature. The accelerated increase in the yield stress and the decrease in the ductility of iron and steels with decrease of temperature are anomalous. A " normal " metal shows not more than 75% increase of yield stress and no decrease of ductility with decrease from room temperature to −188° C.

X-Ray Stress Measurements on Unalloyed Steel When the Yield Point in Compression Is Exceeded. F. Bollenrath and E. Osswald. (Zeitschrift des Vereines deutscher Ingenieure, 1940, vol. 84, July 27, pp. 539–541). It was found by means of X-ray stress determinations that a state of internal stress existed in steel specimens subjected to compressive stresses exceeding the yield point, irrespective of the stress distribution in the elastic state. The internal stresses were concentrated at the core of the specimens. Earlier investigations have shown that the same occurs in tension, but that they are then concentrated at and near the surface.

Cromol Crankshafts. (Automobile Engineer, 1945, vol. 35, Dec., pp. 537–541). The design and advantages of crankshafts made from alloy cast-iron are described. Details are given of the chemical composition and the mechanical properties of Cromol, an alloy cast-iron specially developed for crankshafts. The metal structure that is desirable is described, and some of the important design factors are discussed.

The Effect of the Matrix Structure and Graphite Inclusions on the Properties of Cast Iron. G. Meyersberg. (Iron and Steel Institute, 1946, Translation Series, No. 260). An English translation is presented of a paper which appeared in Teknisk Tidskrift, 1945, vol. 75, June 9, pp. 641–654 (*see* Journ. I. and S.I., 1945, No. II., p. 119 A).

Raised Yield Strength in Bend Tests. G. Brewer. (Metal Progress, 1945, vol. 48, Dec., pp. 1306–1310). It is not generally known that when a solid round or square bar is tested in bending, the stress in the outside fibre may reach a value 40–65% above the limit of proportionality in tension before the bar suffers any permanent set. This is referred to as the "raised yield in bending." To obtain quantitative data on this raised yield strength an investigation was conducted with 2½-in. round bars of a low-alloy chromium-nickel-molybdenum steel, as well as with chromium-molybdenum steel tubes by subjecting them to a pure bending moment and recording the strains by wire-type electrical strain gauges. An account of this investigation is given.

Strain Gage Technique. L. M. Ball. (Proceedings of the Society for Experimental Stress Analysis, 1945, vol. 3, No. 1, pp. 1–20). The practical details of the selection, design, and use of strain gauges are discussed.

Photoelasticity as a Designer's Tool. R. E. Orton. (Proceedings of the Society for Experimental Stress Analysis, 1945, vol. 3, No. 1, pp. 32–39). Examples of the application of photo-elastic stress analysis are described which dispel the impression that this method of investigation involves very expensive equipment and highly trained specialists to operate it.

The Use of a Plastic Model in Crankshaft Analysis. D. M. Pearce and J. Modrovsky. (Proceedings of the Society for Experimental Stress Analysis, 1945, vol. 3, No. 1, pp. 81–93). An experimental method of determining crankshaft main-bearing loads and stresses is described wherein the model is made of plastic to permit the use of smaller applied loads and simplified test equipment.

Stress Analysis of Shafting Exemplified by Saint-Venant's Torsion Problem. T. J. Higgins. (Proceedings of the Society for Experimental Stress Analysis, 1945, vol. 3, No. 1, pp. 94–101). A review is presented of the literature on procedures, curves, formulæ and test data for calculating the stress distribution and maximum stress in filleted, stepped, collared, grooved, slit, and notched shafting.

Dynamic Stress Analysis with Brittle Coatings. G. Ellis and F. B. Stern, jun. (Proceedings of the Society for Experimental Stress Analysis, 1945, vol. 3, No. 1, pp. 102–111). Examples are given of the use of brittle coatings for determining stresses under dynamic conditions. The examples include the stress on a shot-gun during firing and the analysis of the failure of aircraft landing gear.

The Structural Analysis and Significance of Rivet Shear Tests. G. E. Holback. (Proceedings of the Society for Experimental Stress Analysis, 1945, vol. 3, No. 1, pp. 131–154). A simple and effective method of analysing shear-test data for rivets is presented. The non-dimensional nature of this analysis also provides the basis for an analytical approach to the prediction of the structural characteristics of the rivet.

Mild Steel and its Properties, with Special Reference to Its Working and Age-Hardening. C. A. Edwards. (Institute of Vitreous Enamellers: Foundry Trade Journal, 1946, vol. 78, Jan. 24, pp. 83–84). The rôle which titanium plays in ordinary low-carbon steels is discussed. Titanium steel shows no yield point when strained, and does not display any tendency to strain-age-hardening.

X-Ray Stress Measurements at the Beginning of Flow in Carbon Steel. R. Glocker and H. Hasenmaier. (Zeitschrift des Vereines deutscher Ingenieure, 1940, vol. 84, Oct. 26, pp. 825–828). Measurements of the internal stresses in low-carbon unalloyed steels were made using cobalt and chromium radiation, which have different penetrations. It was found that flow began in a 0·01-mm. surface layer at a stress which was only 50–66% of the elastic limit; this applied both in tensile and torsional tests.

The Effect of Periods of Rest on the Time-Resistance and Fatigue Strength of Metallic Materials. F. Bollenrath and H. Cornelius. (Zeitschrift des Vereines deutscher Ingenieure, 1940, vol. 84, May 4, pp. 295–299). Tensile fatigue tests were carried out on specimens of a number of metals and alloys representing a wide variety of lattice structures and mechanical properties, and the effects of periods of rest on the ability of the material to resist a limited number of reversals at above the fatigue strength (*i.e.*, the time-resistance). The test data indicated that rest periods had no effect on the Wöhler curves.

The Effect of Overstress in Fatigue on the Endurance Life of Steel. J. B. Kommers. (American Society for Testing Materials, 1945, Preprint No. 23). The results of an investigation of the effects of overstress in fatigue tests are reported, and these indicate that, in general, when a high overstress has produced a certain percentage of damage to the endurance life, a subsequent lower overstress will show a greater percentage of damage. In the latter case the damage

may be four times greater at the final stress than at the initial stress. On the other hand, when a low overstress is followed by a higher one, the damage to the endurance life is less at the final stress than at the initial one. When the initial overstress is low and the number of cycles not too large, the effect of the initial stress is to increase the endurance life at the final stress. This increase may exceed 100%.

A Sulfur Print Method for the Study of Crack Growth in the Corrosion-Fatigue of Metals. R. C. Brumfield. (American Society for Testing Materials, 1945, Preprint No. 28). A sulphur-printing technique and its application to study corrosion-fatigue damage in specimens of a chromium-molybdenum steel are described.

Work Hardened Surfaces of Fatigue Specimens. F. C. Hull and H. R. Welton. (Metal Progress, 1945, vol. 48, Dec., pp. 1287–1289). A brief account is given of an investigation of the work-hardening effect produced by rough-machining and fine-machining in the preparation of fatigue-test specimens. Rough-machining hardens the material to a greater extent than fine cuts, and even fine-machining hardens steel to a depth of several thousandths of an inch.

X-Ray Diffraction Applied to Shot-Peened Surfaces. E. W. Milburn. (Metal Treatment, 1945, vol. 12, Winter Issue, pp. 259–260). It is well known that the creation of residual compressive stresses by shot-peening the surface of steel will greatly increase its resistance to fatigue. This treatment, however, produces a rough surface; the question therefore arises whether the rough surface can be removed and the high compressive stresses retained. An investigation of the changes in the compressive stress with increasing depth from the surface of a 4-in. shot-peened beam of chromium-molybdenum steel was carried out by the X-ray diffraction technique. The stresses introduced by the shot-peening treatment extended to approximately 0·009 in. A layer 0·005 in. thick could be removed from the surface without diminishing the effectiveness of the shot-peening by more than a few per cent.

A Dynamic Pull-Push Testing Machine for Determining Fatigue Strength and the Damping Capacity of Materials. M. Russenberger. (Schweizer Archiv, 1945, vol. 11, Feb., pp. 33–42). A detailed description is given of a high-frequency electro-magnetic fatigue-testing machine with which specimens can be loaded at up to 1000 kg. at between 50 and 200 reversals/sec. Fatigue-test data obtained with it are presented and discussed and the application of the machine to study damping capacity is described.

Damping Capacity. L. Rotherham. (Metal Treatment, 1945, vol. 12, Winter Issue, pp. 215–222, 232). The engineering and physical aspects of damping are considered and experimental methods for its determination are described. The effects of the carbon and chromium contents of steel as well as the influence of cold-work and the magnetic properties on damping are dealt with.

Microhardness Testing. E. O. Bernhardt. (Zeitschrift des Vereines deutscher Ingenieure, 1940, vol. 84, Sept. 28, pp. 733–736). The construction of the Zeiss microhardness testing instrument is described and its application to study the increase in hardness of cold-worked surfaces is discussed with examples which include an examination of the hardness of shot-blasted wire for valve springs and of rolled threads on chromium-vanadium steel bolts.

Hardness Testing. (Iron and Steel, 1946, vol. 19, Jan., p. 2). A new hydraulically operated semi-automatic machine for making Brinell hardness tests is described.

Testing Depth of Hardness. L. Sanderson. (Metal Treatment, 1945, vol. 12, Winter Issue, pp. 272–274). The use of hardenability in steel specifications is discussed and details are given of the A.S.T.M. standardized procedure for carrying out the Jominy end-quench test.

Remarks on Structural Hardening. A. Portevin. (Revue de Métallurgie, Mémoires, 1944, vol. 41, Feb., pp. 42–44). According to the author's definition the term structural hardening means the increase of Brinell hardness resulting from a modification of the micrographic structure only, *e.g.*, from the heat-treatment which produces troostite or sorbite. On the other hand, the martensitic hardening of steel, which causes a change in the nature of the constituents and in the crystalline network, the author classifies as physico-chemical hardening.

The Mechanism of Brittle Fracture. G. S. Smith. (Metallurgia, 1945, vol. 33, Dec., pp. 55–58). Evidence of the actual process of fracture of brittle materials is not easy to obtain, but recent study of the subject has introduced a new conception of fracture which assists the acceptance of a theory developed 25 years ago by A. A. Griffiths. Particular attention is directed to recent work the results of which appear to establish, with certain kinds of materials, the rôle of external and internal defects during fracture and the order of formation of the parts of the new surfaces.

Contribution to a Study of Brittleness. J. Pomey. (Revue de Métallurgie, Mémoires, 1944, vol. 41, Jan., pp. 17–25; Feb., pp. 49–59; Mar., pp. 83–94). The author reports on a comprehensive investigation of impact strength carried out in the Research Department at the Renault Works in France. Special attention is paid to the impact value in the transverse direction, and the relation between fatigue limit and hardness with reference to the tempering temperature is discussed. A comparison is made between "Uf" and Mesnager specimens for impact testing. The position of the notch, which may be perpendicular or parallel to the grain flow, was examined and the best results were obtained with test-pieces notched perpendicularly to the grain flow. In the last section of the article the author, having proved that the classic methods of impact testing are inadequate for the determination of brittleness, shows that a method can be evolved which, by changing the sharpness of the notch, gives, by extrapolation, the technical cohesive strength of the material.

A Case of Embrittlement of Steel. (Monthly Review of the American Electroplaters' Society, 1945, vol. 32, Aug., pp. 784–786, 845). A zinc-plated spring clip exhibited a stained fracture which was found to be partly intercrystalline in areas adjacent

to spot welds. This condition may be due to local hydrogen embrittlement during plating which in turn may be due to occlusion of hydrogen in microscopic intercrystalline cracks produced in or near the welds during welding. It is probable that clips with sound non-martensitic welds, which might be obtained by preheating prior to spot-welding, would withstand zinc plating without cracking.

Temper Brittleness. J. H. Holloman. (American Society for Metals, Oct., 1945, Preprint No. 11). Recent literature on temper-brittleness is critically reviewed. It appears that this phenomenon is caused by a precipitation from α-iron which occurs preferentially at the grain boundaries. The assumption that the precipitate is a phase consisting of iron and an element which supersaturates α-iron at low temperatures is consistent with the known facts. Above about 600° C. the solubility of the element in steels responsible for temper-brittleness is such that the embrittling precipitate redissolves. At temperatures just below the solubility limit the precipitation takes place very slowly, for the rates of both nucleation and growth are small.

The Magnetic Hardness Testing of High-Speed Steels. H. Springer. (Zeitschrift des Vereines deutscher Ingenieure, 1940, vol. 84, June 1, pp. 365–370). The relation between the magnetic permeability of steel and its hardness is discussed and an instrument is described with which the changes in permeability at each stage of heat-treatment can be followed. The application of this principle to the development of a heat-treatment for high-speed steel is considered.

The Effect of the Field Distribution on the Detectability of Flaws in the Magnetic-Powder Method of Testing. E. A. W. Müller. (Zeitschrift des Vereines deutscher Ingenieure, 1940, vol. 84, July 6, pp. 472–476). The magnetic-powder testing of crankshafts is discussed with special reference to the changes in technique when testing short and long shafts with marked variations in section between the wrist pins and the webs.

Crack Detection—Metrolux Fluorescent Equipment. (Iron and Steel, 1945, vol. 18, Dec., p. 634). The procedure for detecting cracks on a metal surface by impregnating the specimen with a fluorescent compound and then cleaning, drying, and inspecting it under ultra-violet light is described.

The Brittleness and Toughness of Metals at High Temperatures. W. Siegfried. (Schweizer Archiv, 1945, vol. 11, Jan., pp. 1–16; Feb., pp. 43–61). An earlier investigation in the laboratories of Gebrüder Sulzer led to the conclusion that the rapid decrease in the strength of notched specimens and the brittle intercrystalline fractures obtained in unnotched specimens after long periods under load are a result of the same internal phenomena. Two extensive series of creep tests are now described. Tin-cadmium alloys were used in the first series because they behave at room temperature in the same way that steel does at high temperature; cold-drawn as well as heat-treated specimens, notched and unnotched, were tested to examine the effect of the microstructure. In the second series, notched steel specimens were prepared with twelve different notch contours so as to study the effect of the notch angle and the radius at the bottom of the notch. The results proved that the embrittlement followed physical laws which embrace not only low-alloy steels at 500° C., but also the tin alloys. Additional tests at 650° C. on 30/15 and 12/15 austenitic chromium-nickel steels proved that the results with the softer materials were also applicable to austenitic steels.

A method of determining the behaviour of steels at high temperatures which does not involve creep tests lasting several thousand hours is described. Creep tests are made on unnotched specimens; the total elongation at fracture and the reduction in area are measured; these values are used to calculate the uniform elongation and the local elongation by Kuntze's formula. The necking elongation divided by the uniform elongation is thus a measure of the cohesion of the steel.

Structural Changes in Carbon and Molybdenum Steels during Prolonged Heating at 900° to 1100° F., as Affected by Deoxidation Practice. G. V. Smith, R. F. Miller, and C. O. Tarr. (American Society for Testing Materials, 1945, Preprint No. 19a). Annealed and normalized specimens of nine steels (three with carbon 0·1% and six with carbon 0·1–0·2% and molybdenum 0·5%), prepared with different deoxidation practices, were given hardness tests and their microstructures were examined after holding at 900°, 1000°, and 1100° F. for periods up to 5000 hr. The hardness of most of the carbon steels at first increased slightly with time at temperature and then decreased slowly; this behaviour appeared to be unaffected by the deoxidation practice used. The increase was greatest in the annealed samples held at 1000° F. The hardness of the annealed molybdenum steels followed the general pattern of the carbon steels, but the normalized specimens behaved in a manner characteristic of precipitation-hardening systems.

Wrought Heat Resisting Alloys for Gas Turbine Service. C. T. Evans, jun. (Metal Progress, 1945, vol. 48, Nov., pp. 1083–1095). The metallurgy involved in the development and selection of alloys for service at high temperatures, especially for gas turbines, is examined. If premature failure is to be avoided, numerous and complex factors must be taken into account. Short-time tensile, creep and fatigue test data, physical constants, and information on oxidation resistance on 15 of the older alloy steels are presented and discussed.

German Stainless Steels. A. L. Feild. (Iron Age, 1945, vol. 156, Dec. 20, pp. 60–67). Data on the properties, applications and methods of processing and heat-treating the various types of stainless and heat-resisting alloy steels which were commercially produced in Germany during the war, are presented. Scarcity of alloying agents such as molybdenum, nickel and niobium led to the development of low-alloy heat-resistant stainless steels or to the replacement of such elements with manganese, titanium and vanadium. Information on the properties of the steels is presented in seven comprehensive tables.

The Heat-Treatment of Low Alloy Wrought Steels. (Metallurgia, 1945, vol. 33, Dec., pp. 73–78). Particulars are given of the properties which can be

developed by applying suitable heat-treatments to several of the "En" low-alloy steels in the British Standards 970 and 971.

Nickel- and Molybdenum-Free Constructional Steels. H. Kiessler. (Zeitschrift des Vereines deutscher Ingenieure, 1940, vol. 84, June 8, pp. 385–392). The properties and heat-treatment of low-alloy chromium-manganese steels which have been developed in Germany are discussed.

Alloy Steel or Alloy-Treated Steel (?). H. T. Chandler. (Metal Progress, 1945, vol. 48, Nov., pp. 1104–1108). The author differentiates between steel containing alloying elements the purpose of which is to impart certain properties to the steel, and those treated by the addition of agents containing elements for controlling the deleterious effects of sulphur, phosphorus and nitrogen, as well as the grain size. He suggests that carbon steels and the standard alloy steels be specified by their chemical analysis, but that no effort be made to specify a chemical analysis for the residual amounts of addition agents in "alloy-treated" steels which is the term recommended for the second of the above classes of steel.

Wear Resistant Steel Castings for the Mining Industry. J. E. Norman. (Metallurgia, 1945, vol. 33, Dec., pp. 83–84). Most of the wear-resistant castings used by the mining industry are made from alloy steel, either of the austenitic Hadfield manganese type or of one of the lower alloy types in which the composition and heat-treatment are designed for the purpose. The low-alloy steels are favoured by the mining industry as they offer more advantages than austenitic manganese steels. Better machinability and wear resistance, high yield strengths, ease of production, desirable magnetic properties and low first cost are some of the advantages of the former. A brief summary is given of certain types of chromium-molybdenum low-alloy steels which have shown outstanding merit in the mining industry.

Fracture Testing of Alloy Steels for Aircraft Engine Forgings. R. D. Haworth, jun., and A. F. Christian. (American Society for Testing Materials, 1945, Preprint No. A4). Fractures of highly stressed aircraft engine parts have exhibited an unusual appearance characterized by large grains or facets. Photographs of fractures after various heat-treatments are shown and a complete set of fracture-test standards are presented. The operation of the test in production, and results obtained with it, are described.

A Critical Survey of Steel Specifications. R. A. Hoskins. (Australian Institute of Metals: Australasian Engineer, 1945, Nov. 7, pp. 79–85). Steel specifications in England, Australia, and the United States are reviewed. In England the efforts to correct the over-diversification apparent in the pre-war years have been reasonably successful. The British Standard 970-En Series is the most outstanding set which has been evolved, and it is hoped that the ranges of analyses and mechanical properties employed therein will form the basis of most of the post-war general engineering, automotive and aircraft specifications, both in Great Britain and Australia.

METALLOGRAPHY AND CONSTITUTION

(Continued from pp. 31 A–32 A)

Etching for the Microscope. Part I. Iron and Steel. C. A. E. Wilkins. (Metal Treatment, 1945, vol. 12, Winter Issue, pp. 233–241). The purposes of etching and the methods employed are dealt with and the etching of cast iron is then discussed, numerous etching reagents being recommended. A table is included giving heat-tints on grey cast iron at various temperatures. The etching of stainless and high-speed types of steel is also considered.

Continuous X-Ray Inspection. D. Goodman. (Canadian Metals and Metallurgical Industries, 1945, vol. 8, Nov., pp. 32–35, 52). A description is given of the system of continuous X-ray inspection which was installed at one of the largest munition plants in the United States for examining large bombs and shells. A 2000-kV. X-ray unit is employed. The portion of the X-ray beam which is reflected from the anode of the tube is used for the radiography of shells which are carried round the unit on a continuous ring conveyor. The screening and safety methods employed are also described.

X-Ray Diffraction in Inorganic Chemistry, Metallurgy and Mineralogy. H. Lipson. (Nature, 1946, vol. 157, Feb. 2, pp. 124–126). This review of the applications of X-ray diffraction is based on papers presented by A. F. Wells, W. H. Taylor, and the author at a meeting held recently at the Royal Institution. The fundamental importance of X-ray diffraction lies in the insight it gives into atomic arrangements. Work on nickel-aluminium and copper-aluminium alloys is referred to. Many interesting types of diffraction effects have been observed in the study of silicates and these throw light on the mechanism of isomorphous displacement.

Accurate Interpretation of Radiographs. H. R. Clauser. (Engineering Materials and Processing Methods, 1945, vol. 22, Nov., pp. 1418–1422). Definitions are given of a number of terms used in the interpretation of radiographs and the conditions necessary for viewing radiographs are described.

Estimation of Spatial Grain Size. W. A. Johnson. (Metal Progress, 1946, vol. 49, Jan., pp. 87–92). Quantitative relationships may exist between the grain size and properties of a metal or alloy, but their determination is contingent on having more precise means of evaluating grain size than those now available. In this paper details are given of a method of determining the distribution of "spatial" grain sizes from the relative areas of each size of the grains shown in a micrograph. The method is based on the results of statistical observations of structure by E. Scheil and A. Lange-Weise (*see* Journ. I. and S.I., 1937, No. II., p. 247 A).

Effect of Cooling Transformation upon Subsequent Isothermal Reactions. C. A. Liedholm and W. C. Coons. (Metal Progress, 1946, vol. 49, Jan., pp. 104–

107). The importance of paying adequate attention to the differences between the time-temperature cycle of a part to be heat-treated and the time-temperature cycles whereon the S-curve for the particular steel is based is stressed, and two examples are cited in which the structures obtained were not those predicted from the S-curves.

CORROSION OF IRON AND STEEL

(Continued from p. 33 A)

The Corrosion of Iron and Steel. J. C. Hudson. (Journal of Scientific Instruments, 1945, vol. 22, Dec., pp. 231–235). A broad outline of the work and activities of the Corrosion Committee (a Joint Committee of The Iron and Steel Institute and The British Iron and Steel Research Association) is presented.

Recent Russian Work on Corrosion. (Conference on Corrosion, Moscow, 1943: Chemical Age, 1946, vol. 54, Jan. 5, pp. 5–8). A condensed version is given of a paper by T. A. Adjemyau on the corrosion resistance of different materials employed in the construction of plant for Russian chemical industries. It deals with the production of halogen salts such as chlorides, sulphur compounds, fluorine compounds, halogen compounds of the iodo-bromine type, and general research on materials liable to corrode.

A Theory of the Mechanism of Rusting of Low Alloy Steels in the Atmosphere. H. R. Copson. (American Society for Testing Materials, 1945, Preprint No. 25). Exposure tests were started in 1941 on low-alloy steels in industrial and marine atmospheres. Analyses of samples of rust, weather data, and losses in weight are considered. It is postulated that the rate of corrosion depends on the quality and quantity of the water reaching the surface. Relatively insoluble corrosion products decrease the amount of soluble constituents in the water and thereby decrease the corrosion. Relatively insoluble corrosion products do not wash away, so that the porosity of the rust coating and the quantity of water reaching the steel are decreased. In industrial atmospheres, copper and nickel in the steel render sulphate corrosion products more insoluble by forming complex basic sulphates. On mild steel, sulphates in the rust are relatively soluble and promote corrosion, but are washed away by rain. On alloy steels the sulphates are less soluble so that corrosion is slower, but less sulphate is washed away and more accumulates in the rust. The percentage of sulphates in the rust increases as the loss in weight decreases. It has been tacitly assumed that, in a given atmosphere, each test specimen of steel is exposed to the same environment; actually, owing to the influence of corrosion products, each specimen is exposed to a separate environment.

Inhibition of the Corrosion of Metals in Contact with Water and/or Steam. W. Murray. (Association of British Chemical Manufacturers: Steam Engineer, 1946, vol. 15, Jan., pp. 117–118). Recommendations on methods of reducing the corrosion of metals by water and steam are made. Keeping up the *p*H value of the water and adding tannin is advocated for the prevention of "graphitic wastage" in cast-iron economizers.

Controlling Factors in Atmospheric and Immersed Corrosion. W. H. J. Vernon. (Journal of Scientific Instruments, 1945, vol. 22, Dec., pp. 226–230). Factors are discussed which, although apparently subsidiary, are responsible for controlling the rate of corrosive attack under a given set of conditions. The hydrogen-evolution and the oxygen-absorption types of immersed corrosion are dealt with. In the former type it is rare to have an obstructive film at the anode; most frequently the obstruction occurs at the cathode and is associated with the phenomenon of hydrogen overpotential. In the latter type there are two well-known controlling factors, namely, an obstructive film at the anode, and the rate of supply of oxygen to the cathode. In atmospheric corrosion the frequently encountered controlling factor, far from being the rate of oxygen supply, is the rate of supply of water. In the case of iron and steel the controlling factor at high relative humidities is pro vided entirely by the state of purity of the atmosphere. The growth of air-formed oxide films and the significance of rectilinear, parabolic, logarithmic, and asymptotic oxidation-time curves are considered.

Electrochemical Measurements for Corrosion Studies. P. T. Gilbert. (Journal of Scientific Instruments, 1945, vol. 22, Dec., pp. 235–237). A detailed description is given of an apparatus which was set up to study the behaviour of pairs of coupled metals in tap water to establish the relationships between them as they corroded with different concentrations of dissolved oxygen and carbon dioxide in the water. The potentials and corrosion currents from four couples of corroding metals were simultaneously and continuously recorded.

The Corrosive Action of Benzole Absorption Oils. (Coke and Smokeless Fuel Age, 1946, vol. 8, Jan., pp. 17–18). Two reports of independent research work on the corrosive action of benzole absorption-oils have recently been published, one by O. B. Wilson, and the other by C. M. Cawley and H. E. Newall (*see* Journ. I. and S.I., 1945, No. II., p. 69 A and 1946, No. I., p. 33 A). The conclusions reached are compared and discussed in the present paper.

Intercrystalline Cracking. C. D. Weir. (Mechanical Engineering, 1945, vol. 67, Dec., pp. 834–835). The effect of the silica/sodium-hydroxide ratio in the feed water on the susceptibility of boiler steel to cracking is treated statistically and diagrams showing the boiler conditions (*i.e.*, pressure, sodium-hydroxide concentration, and the above ratio), in relation to the cracking of boiler plates are presented. It is concluded that this ratio does not influence the frequency with which intercrystalline cracking occurs.

The Behaviour of Zinc-Iron Couples in Carbonate Soil. J. H. Gilbert and G. Corfield. (Corrosion, 1945,

vol. 1, Dec., pp. 187–191). The behaviour of zinc in contact with steel in soils, important from the standpoint of the cathodic protection of pipe lines, is described. Each couple consists of an iron ring connected to one or more zinc cylinders, and the entire surfaces of the iron rings were in contact with a "high-carbonate" soil near Los Angeles for 546 days. The variations in current, the electrode potentials, and the loss in weight of the metals were determined.

Marine Corrosion Testing Station at Kure Beach, North Carolina. (Corrosion, 1945, vol. 1, Dec., pp. 178–186). The sea water corrosion-testing station at Kure Beach is described, with numerous illustrations (*see* Journ. I. and S.I., 1945, No. II., p. 150 A).

Recommendations on Methods of Protection against Corrosion for Light Gauge Steel and Wrought Iron Used in Permanent Building Construction. (British Standards Institution, 1945, P.D. 420). This pamphlet is not a specification, but is to be regarded as a reasoned statement of recommended procedure for the protection of light gauge wrought iron or steel sections or sheets as used in building construction, more particularly for dwelling houses.

BOOK NOTICES

(Continued from p. 35 A)

BARR, W., and A. J. K. HONEYMAN. "*Steel and its Practical Applications.*" Second Edition. 8vo, pp. xii + 156. Illustrated. London and Glasgow, 1945: Blackie and Son, Ltd. (Price 8*s.* 6*d.*)

Since the first edition of this book, considerable changes and advances have been made in ferrous metallurgy. The war has been responsible especially for a changed outlook regarding alloy steels, hardenability, weldability and creep. This second edition, which contains brief accounts of these modern trends, should be welcomed by engineers. The chapters dealing with carbon and alloy steels have been brought up-to-date by the inclusion of details of the new "En" steels. It has been necessary to enlarge the chapter on heat-treatment to include reference to isothermal-transformation (or S) curves, and the Jominy test for hardenability.

New chapters have been added dealing with the factors influencing the creep strength of steel and metallurgical aspects of gas-cutting and welding high-tensile steels. The illustrations are of good quality, but it would facilitate reading if the position of the plates was near the text.

E. C. ROLLASON.

ROLFE, R. T. "*A Dictionary of Metallography.*" 8vo, pp. viii + 243. London, 1945: Chapman & Hall, Ltd. (Price 15*s.*)

The evolution of any branch of physical science inevitably calls into existence a specialized terminology, in which familar words may have circumscribed meanings, and new words may have been coined to describe new facts and phenomena. Metallography had a flying start as regards nomenclature, for almost the first terms used in the infancy of the science were the personal inventions of the pioneer metallographists, giving us the now familiar names for the microscopic constituents of steel. Metallography now is, as the author of this dictionary points out in his foreword, a division of metallurgical science difficult to define, as it undoubtedly reaches out and touches the fringes of many other sciences, so that the necessary vocabulary includes many terms in general scientific use, and thus an exhaustive dictionary of the subject would be indeed a formidable work. The author makes no claim to exhaustiveness in this sense, and having resigned himself to certain deliberate omissions, has covered the remaining field in a thorough manner. It may at once be said that no serious omission of any metallurgical term was noted, and the fairly generous treatment given to the vocabulary of most of the ancillary sciences should satisfy the great majority of enquirers.

Indeed, so wide is the net cast by the compiler that the reviewer seriously questions the accuracy of his title, notwithstanding the plea advanced in the foreword for a generous interpretation of the term "metallography." Premeditated omissions can readily be accepted if the very best use is made of the available space, but it is debatable whether the author is justified in allowing some descriptions of manufacturing processes and data on the properties, occurrence, and extraction of the metallic elements to displace all reference to, for example, the magnetic properties of metals or microscope optics.

Every lexicographer, from Dr. Johnson onwards, has had to face the heavy task of maintaining unrelenting accuracy. A dictionary is nothing if not reliable, and Mr. Rolfe comes very well out of the most exacting factual enquiry that the reviewer could apply. It is therefore with due deference to the predominating success achieved by the author that some lapses are mentioned. Thus, the well-known S.80 composition of stainless steel is included under the term "Stainless Iron," a classification that is at variance with its properties and with the general position accorded to this alloy in specifications and normal engineering parlance. Again, the statement "In their shorter reaches, ultrasonic waves merge with radio and finally with light waves" is made under "ultrasonics." Neither sonic nor ultrasonic waves have any place in the electromagnetic wave spectrum. Lastly, the author, under "Charpy Impact Test" rather inexplicably cites the Addendum of March 1942 to British Standard 131–1933 as referring to Charpy test-pieces with the Charpy keyhole notch, whereas the publication mentioned specifies the Mesnager type notch and the two are not interchangeable.

The book is well bound and clearly printed, the item headings standing out well on each page. The alphabetical arrangement naturally provides its own index, and cross indexing is freely provided when synonymous terms are defined. The author makes use of standard definitions where possible, mainly from American sources, many of these relating to the growing technology of powder metallurgy. The reviewer found rather distracting the practice of using only the initial letters of the item headings, when the same words recur in the text, although this is common in ordinary dictionaries, but surely such compression is carried too far, when nickel is rendered as "n" in place of the universally acceptable "Ni."

S. L. ROBERTON.

REFRACTORY MATERIALS

Steel Plant Refractories. J. H. Chesters. (British Engineering Export Journal, 1946, vol. 28, Jan., pp. 428–433). Recent improvements in the manufacture of steel-plant refractories, including the control of quality by statistical methods, are pointed out. Methods of testing bricks and the properties of magnesite and carbon refractories are discussed.

Potmaking. (British Steelmaker, 1946, vol. 12, Jan., pp. 24–25). A detailed description is given of the production of crucibles for steelmaking.

Recent Progress in the Manufacture of Refractories for Metallurgy. Y. Lefort. (Revue de Métallurgie, Mémoires, 1942, vol. 39, May, pp. 141–151; June, pp. 172–180; July, pp. 201–210). Three main groups of refractories, i.e. the alumina, silica and magnesite types, are discussed in detail. Their nature, structure and behaviour at high temperatures are examined, and the progress in their manufacture reviewed. In conclusion the importance is stressed of scientific investigations, such as thermal analysis, microscopical examination and X-ray testing, for the development and improvement of refractories.

Carbon Lining for Blast Furnaces. F. G. Vosburgh. (Steel, 1945, vol. 117, Dec. 24, pp. 86–90; Dec. 31, pp. 62–63, 78–80). The history of the use of carbon linings for blast-furnace hearths in the United States is outlined, and descriptions are given of the blocks used, the positions of the joints, and the carbon ramming mix.

Statistical Treatment of Laboratory Test Data on Refractories. J. H. Chesters and I. M. D. Halliday. (Transactions of the British Ceramic Society, 1945, vol. 44, Sept., pp. 135–166). Research on refractories is passing from a qualitative to a quantitative stage. Variability, both in the product and in operating conditions must be measured and allowed for. This requires the application of statistics. The use of histograms, in which the frequency with which particular values of a property occur is indicated by vertical strips, often gives a qualitative answer, but the calculation of the arithmetic mean, the standard deviation and the range are often used in support. Quality control charts are shown to provide a rapid method of deciding whether the product under test differs significantly in properties from that received previously. It is also shown that arithmetic probability paper affords a ready means of comparing data of the normal distribution type, particularly as it enables quick estimates to be made of the rejections that would occur at various levels of control. It is shown that apparent differences between the quality of dolomite received during two periods may be due to inadequate sampling. The use of correlation coefficient is illustrated with reference to the close relationship between the bulk density and thermal conductivity of diatomite bricks.

Statistical Data Relative to the Firing Expansion, Specific Gravity and After-Expansion of Silica Bricks. T. R. Lynam and I. M. D. Halliday. (Transactions of the British Ceramic Society, 1945, vol. 44, Oct., pp. 167–202). It is regarded as unsatisfactory that the quality of a wagon-load of silica bricks for the gas industry is judged by destructive tests on two bricks selected at random. If a definite relationship could be proved to exist between the firing expansion (linear) which occurs in the kiln and the specific gravity of the fired product, the simple expedient of measuring a brick should be an efficient test of the heat-treatment it has received. It would then be possible to apply this rapid test to a much larger number of samples. An account is given of a statistical investigation of test data on 200 random samples from different known positions in various kilns. The tests included kiln expansion (linear), specific gravity, and after-expansion. It was found that a significant relationship exists between the kiln-firing expansions and the specific gravities of silica products, and that the non-destructive measuring test can be applied to ascertain the heat-treatment.

Testing of Chrome-Magnesite Brick for Resistance to Iron Oxide Bursting. S. Zerfoss and H. M. Davis. (Journal of the American Ceramic Society, 1946, vol. 29, Jan., pp. 15–20). The probable mechanism of the iron-oxide bursting of chromium-magnesite brick is discussed. An attempt was made to duplicate this in the laboratory. The results indicated that the bursting can be simulated in a laboratory experiment, providing the atmosphere of the furnace is such that the penetrating iron oxide liquid is rich in magnetite. A test is proposed for characterizing the bursting resistance of various bricks.

X-Ray Diffraction Data for Compounds in the System CaO–MgO–SiO_2. C. B. Clark. (Journal of the American Ceramic Society, 1946, vol. 29, Jan., pp. 25–30). X-ray diffraction data for compounds in the system CaO–MgO–SiO_2 obtained by the powder method are widely scattered in the literature. In certain instances there is need of revision because of improvement in equipment and technique. X-ray powder photographs have therefore been made, and data are presented in this paper on nineteen compounds in this system. These represent all the known compounds which are stable, and several which are metastable, at room temperature.

FUEL

(Continued from pp. 37 A–38 A)

Problems in Fuel Efficiency. C. Hulse and R. J. Sarjant. (Iron and Steel Institute, 1946, this Journal, Section I.). Problems in fuel efficiency relating to works where processes may be specialized are discussed in relation to the bearing of fuel selection on efficiency, developments in furnace design and

construction, methods of fuel saving in individual works, education, and research. The present position is reviewed and lines of future development are suggested.

Heat Transfer. G. T. Skaperdas. (Industrial and Engineering Chemistry, Industrial Edition, 1946, vol. 38, Jan., pp. 10–11, 32, 33). Progress made during the last two years in the theory and practice of heat transmission are reviewed. There are 75 references to the literature.

The Influence of Tube Shape on Heat-Transfer Coefficients in Air to Air Heat Exchangers. F. H. Green and L. S. King. (Transactions of the A.S.M.E., 1946, vol. 68, Feb., pp. 115–122). Heat-transfer coefficients are reported for air passing through round tubes, through partially flattened tubes, and through partially flattened tubes which have been dimpled. It was found that flattened tubes maintained fully turbulent flow, and that they therefore had a higher heat-transfer coefficient at lower Reynolds numbers than round tubes. The addition of dimples to the flattened tubes maintained fully turbulent flow from the transition point and increased the heat-transfer coefficient substantially. The addition of the dimples also improved the relation between the heat transfer and the pressure drop in the flattened tubes.

Problem of Flame Radiation, with Particular Reference to Melting Furnaces. T. F. Pearson. (Iron and Coal Trades Review, 1946, vol. 152, Mar. 1, pp. 381–385). To take the ratio of useful heat output to the total heat input as the measure of the efficiency of a gas-fired furnace is not altogether adequate, since the degree of heat transfer from the flame and walls tends to be overlooked. This detail of the simplified conception of furnace efficiency is of greater importance in high-temperature melting furnaces than in low-temperature heat-treatment furnaces. Reasons for this are given in the present paper and the possibilities of increasing the thermal efficiency by modifying the flame are discussed.

Combustion Control. (Steel, 1945, vol. 117, Dec. 10, pp. 134–139). Three methods are described which are used for combustion analysis in open-hearth furnaces, soaking pits, treating furnaces, steam power plants, and internal combustion engines. In particular, an instrument is described which has two dials; one indicating the percentage of unused combustible gases and the other the percentage of excess oxygen which may be present in the spent gases. These direct readings can be made at any instant from samples taken from any part of the furnace, flue or exhaust.

Conservation of Coal. A. C. Pasini. (Iron and Steel Engineer, 1946, vol. 23, Jan., pp. 112–114). Recommendations are made for obtaining the most efficient combustion of coal and for maintaining boilers in proper working condition.

Tubular Metallic Recuperators. G. N. Critchley. (Institute of Fuel Wartime Bulletin, 1946, Feb., pp. 100–119). A comprehensive discussion of the factors influencing the design and use of metallic recuperators is presented in four parts. The first three parts are of a mathematical nature and give methods of answering the following questions: (1) How much heat does the waste gas contain? (2) How much heat could be returned to the furnace by (*a*) the combustion air, and for (*b*) the fuel? (3) What, in given circumstances, is the most suitable design of recuperator? In the fourth part the economic aspects of recuperation are discussed.

An Experimental Determination of the Factors Governing the Design of Regenerators with Special Reference to Coke Ovens. T. C. Finlayson and A. Taylor. (Journal of the Institute of Fuel, 1946, vol. 19, Feb., pp. 82–98). After outlining the general principles of a regenerator, defining regenerator efficiency and briefly reviewing the literature, a report is presented on an investigation at a specially built test plant. The main object was to find the effect and quantitative value as regards heat transfer and pressure loss of the different factors in the design of both double-staggered checkers and straight-through or any other promising types of fillings, so that it would be possible to calculate the effect of any variations in design, and to indicate the most profitable lines for the development of coke-oven regenerators.

The Economic Limits of Separation of Middlings in Dense Medium Washers. K. F. Tromp. (Journal of the Institute of Fuel, 1946, vol. 19, Feb., pp. 112–115). When the middlings are used as a boiler fuel the separation between the middlings and the shale should be chosen at a density where the middlings have such a high ash content that the cost of industrial firing just equals the amount of its heat contribution. This ultimate allowable ash content, called the "economic firing limit" was originally assessed by Haarmann at 65% when the middlings are fired at the colliery. The author has reinvestigated the various factors involved, and arrives at the conclusion that normally the density cut will be much lower (at about 56% of ash), but that under certain conditions higher ash contents (up to 70%) can still advantageously be fired. Due to the imperfection of coal-washing systems, the actual cut has normally to be located at a density other than that of the economic firing limit, in order to be certain that the economic loss due to migrated material is at a minimum.

The Newcastle Steel Works Cleans Coal. W. H. Wainwright and J. S. Ratcliffe. (B.H.P. Review, 1945, vol. 23, Dec., pp. 6–8). A description is given of the Barvoys coal-cleaning plant at the coke-ovens of the Broken Hill Proprietary Co., Ltd., Newcastle, New South Wales.

Pulverised Coal and the Modern Drop Forge. J. M. Austin. (Institute of Fuel Wartime Bulletin, 1946, Feb., pp. 122–130, 134). A detailed description is given of an installation of fifty furnaces fired by pulverized coal for heating forgings. They include both batch and continuous furnaces. Full details are given of the refractories and insulation used and of the measures taken to reduce slagging and the maintenance costs associated with it. The properties and preparation of the fuel and plant operation and performance are also dealt with.

The Agglutination and Swelling of Coking Coal. H. E. Blayden, H. L. Riley, and F. Shaw. (Fuel in Science and Practice, 1946, vol. 25, Jan.-Feb., pp. 13–24). The literature on agglutinating tests for coal is reviewed and some limitations of the British

Standard Method for determining the agglutinating value are pointed out. Experiments are described in which coke buttons were prepared from coal mixed with colloidal magnesium trisilicate, and crystalline and amorphous varieties of silica and alumina. The factors inherent in the coal itself which affect the strength of the coke button are: (*a*) The proportion and character of the agglutinating constituents, *i.e.*, the coking-coal bitumens; (*b*) an inherent strength factor possibly determined by the pressure on the coal during formation; and (*c*) the extent of the swelling during carbonization. Laboratory tests can provide a valuable guide to the correct choice of carbonizing conditions and proper blending to obtain a coke with optimum properties.

Coke Oven and B. F. Practice. J. H. Patchett. (Cleveland Institution of Engineers: Iron and Steel, 1946, vol. 19, Feb., pp. 63–68). A general survey of the progress in coke-oven and blast-furnace practice in Great Britain in recent years, with indications of the direction which future developments are likely to take, is presented.

Method for Obtaining Ammonium Chloride from Gas-Liquor. C. H. Bone. (Blast Furnace and Steel Plant, 1945, vol. 33, Dec., pp. 1507–1510). A method of recovering ammonium chloride, which is satisfactory for use in galvanizing flux, from the gas-liquor produced during the carbonization of coal is described. The process has been worked successfully on a pilot-plant scale.

PRODUCTION OF IRON

(Continued from p. 38 A)

The Profile of Blast Furnaces. V. Rémond. (Revue de Métallurgie, Mémoires, 1945, vol. 42, Feb., pp. 29–64). A report is presented on a comprehensive study of blast-furnaces. It contains: (1) A table of profiles of 56 French blast-furnaces; (2) a comparison of the dimensions of these blast-furnaces and of data on their operation; (3) data relating to the production and efficiency in various blast-furnace plants, and (4) the author's conclusions presented in the form of a study of blast-furnace parameters. Finally, the evolution of the blast-furnace profiles in France from 1920 to 1940 is described, and data relating to blast pressure are given.

Three Blast-Furnace Questions. O. R. Rice. (Blast Furnace and Coke Oven Association of the Chicago District and Eastern States Blast Furnace and Coke Oven Association: Blast Furnace and Steel Plant, 1946, vol. 33, Dec., pp. 1523–1529). The following three questions relating to blast-furnace operation are discussed: (1) Does easier driving reduce the coke rate per ton of pig-iron? (2) Do little furnaces have lower coke rates than big ones? (3) Does a relatively large working volume conduce to a lower coke rate? Data on these questions are presented and considered, and the following conclusions are reached: (1) Easier driving does seem to reduce the coke rate, but not to a practical or economic degree; (2) small furnaces seem to have a lower coke rate than large ones; and (3) relatively large working volumes appear to result in a lower coke rate.

Blowing Out of Blast-Furnaces. O. Becker. (Blast Furnace and Coke Oven Association of the Chicago District and Eastern States Blast Furnace and Coke Oven Association: Blast Furnace and Steel Plant, 1945, vol. 33, Dec., pp. 1530–1531). **Full Furnace Blowdown Method.** O. Becker. (Steel, 1946, vol. 118, Jan. 14, pp. 112–115). A method of blowing out a blast-furnace is described. The stack is kept full of nut coke which is wetted profusely in the skip. The blast is gradually reduced as the pressure drops, and after the last cast is made, the gas is drafted back, blowpipes are removed and the tuyeres are plugged. The coke may be quenched with water from above or below, but preferably, and with greater safety, steam can be used.

Use of Domestic Coke in Blowing a Blast Furnace. W. H. Burnett. (Blast Furnace and Coke Oven Association of the Chicago District and Eastern States Blast Furnace and Coke Oven Association: Blast Furnace and Steel Plant, 1945, vol. 33, Dec., pp. 1533). **Domestic Coke Blowout Method.** W. H. Burnett. (Steel, 1946, vol. 118, Jan. 14, pp. 109–110). A procedure for blowing out a blast-furnace using domestic coke is described. Clean-out blanks (coke without ore) are charged daily for several days before blowing out, and the limestone in the burden is reduced. When ready, domestic coke only is charged with sufficient water in the skip to keep the top temperature below 600° F. Blast is taken off after the last cast is finished, the top is opened and full steam is turned into the dust catcher. Water may be introduced through the top to cool the stack, in which case it is sufficiently cooled in about 18 hr. Two coolers are removed and a chute is put in position to discharge the water and nut coke into trucks.

Blowing Out a Blast Furnace Utilizing the Nut Coke-Hydraulic Wash-Out Method. J. R. Barnes. (Blast Furnace and Coke Oven Association of the Chicago District and Eastern States Blast Furnace and Coke Oven Association: Blast Furnace and Steel Plant, 1945, vol. 33, Dec., pp. 1535–1536). **Nut Coke Hydraulic Washout Method.** J. R. Barnes. (Steel, 1946, vol. 118, Jan. 14, pp. 110–112). The practice employed by the Republic Steel Corporation for blowing out blast-furnaces is described in detail. After the last cast the furnace is filled with nut coke, and cooling water and steam are injected, the procedure being similar to that described by W. H. Burnett (*see* preceding abstract).

The Krupp-Renn Process. (Iron and Coal Trades Review, 1946, vol. 152, Mar. 15, pp. 485–491). *See* p. 38 A.

Hot-Pressing of Iron Powders. O. H. Henry and J. J. Cordiano. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1919: Metals Technology, 1945, vol. 12, Oct.). An

investigation is reported the purpose of which was to determine the properties of hot-pressed specimens of electrolytic iron powder. Three methods of supplying heat to the specimens during the pressing operation were employed, viz., induction, passing a current through the highly resistant mass of powder, and enclosing the entire die assembly in a furnace. Tensile, density, and hardness tests were conducted on the specimens. There was a decided increase in density with increasing temperature. The curves for tensile strength and elongation clearly show the advantages that may be gained from hot-pressing. The increase in the surfaces of the particles coming into contact and the increase in the diffusion rates with rising temperature under sustained pressure are mainly responsible for the intraparticle bonding.

Pressing Complicated Shapes from Iron Powders. C. G. Goetzel. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1920 : Metals Technology, 1945, vol. 12, Oct.). Two processes for moulding complicated parts of uniform density from powdered metals are described in detail. The first refers to curved parts, and the second is especially adapted to parts having one or more recesses or steps.

FOUNDRY PRACTICE

(Continued from p. 39 A)

Gray Iron Acceptance Broadens. T. Barlow. (Foundry, 1946, vol. 74, Jan., pp. 70–75, 162). Developments in processes and equipment at American grey iron foundries during the war are reviewed.

What Type Charger and Charging Bucket. A. W. Gregg. (Foundry, 1945, vol. 73, Dec., pp. 84–87, 228–232). Factors affecting the selection of mechanical equipment for charging cupolas are discussed and a number of charging-bucket designs are critically reviewed.

Foundries of the Future. J. B. Lamenzo. (Foundry, 1945, vol. 73, Nov., pp. 84–87, 238–246; Dec., pp. 88–93, 258–262). Factors to consider in the installation of conveyor systems and other labour-saving equipment in the foundry are discussed.

Foundry Practice at Consolidated. T. R. Stanley. (Canadian Metals and Metallurgical Industries, 1945, vol. 8, Dec., pp. 24–26). A description is given of the foundry practice at the works of the Consolidated Mining and Smelting Company of Canada, Ltd., British Columbia, where a wide variety of iron, steel, and non-ferrous metal castings are made.

Wartime Foundries of China. E. K. Smith. (Foundry, 1945, vol. 73, Dec., pp. 83, 210–218). Some of the difficulties encountered and overcome during the war by foundries in China are described.

The Manufacture and Properties of Modern High-Strength Castings. Part I. Steel Castings. C. H. Kain. (Manchester Association of Engineers, Feb., 1946, Advance Copy). A general account of the production of medium and light steel castings is given. The properties and methods of testing are discussed, with special reference to the casting of test bars.

The Manufacture and Properties of Modern High-Strength Castings. Part 2. High Duty Cast Irons. J. J. Sheehan. (Manchester Association of Engineers, Feb., 1946, Advance Copy). Progress in the production of high-duty cast irons is reviewed, with notes on the properties of cast-iron valve-tappets, low-chromium irons for wear-resistance and high-silicon irons for heat-resistance. A syphon pouring arrangement for the cupola spout which separates the slag from the metal and creates a substantial head of incandescent slag in the cupola is described.

Some Notes on High-Duty Cast Irons. A. H. Horton. (Institute of British Foundrymen : Foundry Trade Journal, 1946, vol. 78, Feb. 14, pp. 169–171). Methods of making high-duty cast iron in the cupola with subsequent additions of inoculants are described, and the properties of the iron are discussed.

Malleable Industry Improves Plant Facilities. C. H. Lorig. (Foundry, 1946, vol. 74, Jan., pp. 92–96, 214–218). Some of the very modern equipment installed at American malleable iron foundries during the war is described and illustrated.

West Coast Steel Foundry Uses Large Furnaces. R. L. Collier. (Foundry, 1945, vol. 83, Dec., pp. 99–100). A brief description is given of the electric furnaces and equipment which were installed for a new steel foundry at Portland, Oregon, for casting stern frames and other heavy castings for the shipyards.

The Manufacture of Some Thin-Walled Steel Castings, with Notes on the Influence of Pouring Speed. The Foundry Practice Sub-Committee. (Journal of the Iron and Steel Institute, 1945, No. II., pp. 519P–534P). This paper is presented by the Foundry Practice Sub-Committee of the Steel Castings Research Committee. Short accounts are given of the moulding and founding procedures used at seven foundries in the production of thin-walled steel castings, and of the influence of the rate of pouring on the quality of the castings produced. With bomb castings, least difficulty was met with when the moulds were filled quickly, and also less trouble was encountered with hot steel. There was considerable scabbing with very slow casting or with very cool steel. High rates of pouring also reduced the number of rejections due to hot tears, pulls, and other defects, such as sand inclusions. Casting temperatures above 1580° C. gave better results than temperatures below, say, 1550° C. With axle cases, satisfactory castings were made over a wide range of pouring speeds, but it should be noted that the casting temperatures are available for only one foundry.

Steel Castings Find New Applications. C. E. Sims. (Foundry, 1946, vol. 74, Jan., pp. 76–81, 220–225). Wartime developments in the manufacture of steel castings in the United States are reviewed.

Brockhouse—Growth of a Great Polygon Industry. (British Steelmaker, 1946, vol. 12, Jan., pp. 10–16). A brief account (based on Compton Mackenzie's book " Brockhouse ") is given of the life and work of John

Brockhouse of West Bromwich, founder of the Company bearing his name. John Brockhouse became the central figure in a great industrial combination embracing twenty-five individual companies making iron and steel castings, heavy chain, and machinery of all kinds.

1944–45 Sintering Test. (American Foundryman, 1945, vol. 8, Dec., pp. 38–39). This paper is a short report by the Committee on Sintering Test of the American Foundrymen's Association; in it the causes of the " spotty burning on " of sand on a steel casting are examined. This defect is attributed to insufficient ramming.

Effective Use of Cereal Core Binders. A. E. De Clercq and T. Barlow. (Foundry, 1945, vol. 73, Dec., pp. 109, 192–210). The effect of corn-flour on the green strength, dry strength, permeability, baking time and surface hardness of core-sand mixtures is discussed and recommendations on the use of corn-flour as a binder are made.

Dust Removal at Belt Transfer Points. C. C. Hermann. (Foundry, 1945, vol. 73, Dec., pp. 107, 226–228). In mechanized foundries where the sand is conveyed on belts there is always dust at the end of the belt where the sand is tipped off on to another belt. Details are given of the design of hoods to place over the conveyors at these points.

High-Pressure Feeding of Static Molds. S. T. Jazwinski and S. L. Finch. (Iron Age, 1946, vol. 157, Jan. 10, pp. 59–63; Jan. 17, pp. 54–63). *See* p. 20 A.

The Feeding of Castings. R. W. Ruddle. (Institute of Australian Foundrymen: Foundry Trade Journal, 1946, vol. 78, Mar. 7, pp. 253–256). The mechanism of the solidification of metals in moulds is explained and the essential requirements for good feeding practice are pointed out, with notes on the dimensions and position of risers. Finally, the method of applying atmospheric pressure to blind risers by inserting a core of permeable material in the riser cover is described.

Some New Foundry Plant. (Engineer, 1946, vol. 181, Feb. 15, pp. 151–152). Illustrated descriptions are given of new core-blowing, core-making and shot-blasting machines which have been developed by a Birmingham company.

The Formation of Banded Structures in Horizontal Centrifugal Castings. C. Howson. (Foundry Trade Journal, 1946, vol. 78, Feb. 21, pp. 191–199). The formation of banded structures in centrifugal castings has been correlated with the degree of vibration of the casting machine in an earlier paper by L. Northcott and D. McLean (*see* Journ. I. and S. I., 1945, No. I., p. 303 P). In the present paper an effort is made to show that the existence of vibration is not a necessary condition, as in horizontal casting there is an inherent variation in mechanical force during rotation. The first section deals with mathematical and theoretical considerations with regard to centrifugal casting; some examples of ferrous and non-ferrous-metal centrifugal castings are described, and finally an attempt is made to explain the macrostructural formations encountered.

Centrifugally Cast Aircraft Parts. J. F. B. Jackson. (Aircraft Production, 1945, vol. 7, Nov., pp. 511–513). The application of the centrifugal casting process to irregular and asymmetrical units in high-tensile alloy steel is described, and the results of typical mechanical tests on centrifugal castings for aircraft are given.

Foundry Research at Watertown Arsenal. W. G. Gude. (Foundry, 1945, vol. 73, Dec., pp. 104–106, 186–188). A brief description is given of some of the research work on the centrifugal casting of steel gun-barrels which has been carried out at Watertown Arsenal in the United States.

Methods Employed to Obtain Rates of Solidification. K. L. Clark. (American Foundryman, 1945, vol. 8, Dec., pp. 24–25). This paper constitutes the first part of a report on work done by the Heat Transfer Committee of the American Foundrymen's Association. It describes tests in which a series of steel plates $6 \times 6 \times 2$ in., $8 \times 8 \times 4$ in., $12 \times 12 \times 6$ in., and $12 \times 8 \times 4$ in. were cast in dry-sand moulds and in chills. The flasks were overturned at selected time intervals after pouring, and the relationships between the thickness of the solidified skin and time were determined. The results obtained are given in tables, and in another part of the report (*see* next abstract) they are compared with those obtained using the heat-flow analysis developed by Paschkis (*see* Journ. I. and S. I., 1944, No. II., p. 267 P).

Studies on Solidification of Castings. V. Paschkis. (American Foundryman, 1945, vol. 8, Dec., pp. 26–37). This paper is the second part of the report on work done by the Heat Transfer Committee (*see* preceding abstract). Before the electrical analogy method can be used to attack heat-transfer problems in the solidification of castings, proof is required that the method is suitable. Data were therefore obtained by direct methods in the foundry, and these are compared in the present paper with data obtained with the mass-flow analyser. The electrical analogy method appears to be a promising tool, and it appears possible to establish with reasonable approximation one curve for sand casting and one for chill casting, showing the freezing times as functions of metal thickness. This approximation is better for smaller than for greater thicknesses.

PRODUCTION OF STEEL

(Continued from pp. 40 A–42 A)

Steel and the Engineer. A. Roebuck. (Journal of the Junior Institution of Engineers, 1946, vol. 56, Jan., pp. 93–96). The development of steelmaking in England is briefly reviewed, with notes on the work of eminent metallurgists and engineering achievements based on the use of steel.

Developments in the Iron and Steel Industry during 1945. W. H. Burr. (Iron and Steel Engineer, 1946, vol. 23, Jan., pp. 53–65). Progress made in the United States in 1945 in the production and finishing of iron and steel is reviewed.

German Toolsteel and Special Steel Industry. J. P.

Gill. (Metal Progress, 1946, vol. 49, Jan., pp. 122–124). A short account is given of the organization of the German tool-steel and special steel industry, with notes on the manufacturing practices.

The Production of Very Clean Steel in the Basic Bessemer Shop. J. Klärding. (Iron and Steel Institute, 1946, Translation Series, No. 268). This is an English translation of a paper which appeared in Stahl und Eisen, 1945, vol. 65, Mar. 1, pp. 118–121 (*see* Journ. I. and S.I., 1945, No. II., p. 35 A).

The Perrinisation of Basic Bessemer Steel in France. (Metallurgia, 1946, vol. 33, Jan., pp. 115–116). The Perrin steelmaking process, which was described by B. Yaneske (*see* Journ. I. and S.I., 1940, No. II., p. 35 P), underwent further development at Valenciennes in secret during the German occupation in the 1940–44 period, and a brief account of some of the work is given. It was found that practically pure iron could be produced by Perrinising basic-Bessemer steel with new aluminous slags. Normal basic converter steel at the end of the blow is slagged clean and a new basic slag put on. The steel is not over-oxidized. A ladle containing the aluminous slag is brought to a tapping position under the mouth of the vessel. The special slag, previously melted in an electric furnace, consists of lime and alumina; to prevent any phosphoric slag coming over in the tapping operation, lump lime is positioned at the mouth of the vessel to form a bridge. The vessel is then tipped quickly enough to give a rather violent intermixing action of the liquid slag and steel in the ladle. Additions of aluminium or ferrosilicon are made to the stream during tapping. The steel when cast into moulds is completely killed.

A Laboratory Investigation of the Phosphorus Reaction in the Basic Steelmaking Process. K. Balajiva, A. G. Quarrell, and P. Vajragupta. (Iron and Steel Institute, 1946, this Journal, Section I.). A new technique, developed for melting slags and metal together at controlled temperatures and for taking satisfactory samples of slag and metal, is described. The technique has been applied to a study of the phosphorus reaction at 1585° C., using prefused slags containing all the normal constituents except CaF_2, but with a much wider range of composition than is normally encountered in industry. It is shown that the empirical relationships which have previously been developed for purposes of industrial slag control break down when applied to slags having a wide range of compositions. On the other hand, a linear relationship is found to exist between the logarithm of the phosphorus equilibrium constant

$$\left(k = \frac{(P_2O_5)}{[P]^2\,(FeO)^5}\right)$$

and the logarithm of the total lime content of the slag. On the basis of this relationship, curves are drawn to assist in slag control at 1585° C.

X-ray and petrographic examinations of the slags have been carried out and three constituents have been identified: (1) A solid solution of $3CaO.P_2O_5$ in $2CaO.SiO_2$, (2) $2CaO.Fe_2O_3$, possibly in solid solution with $4CaO.Al_2O_3.Fe_2O_3$, and (3) an oxide phase consisting mainly of FeO, MgO, and MnO in solid solution. The free-lime contents of the various slags are calculated on the basis of $3CaO.P_2O_5$ and $2CaO.SiO_2$ and various assumptions with regard to Fe_2O_3 and Al_2O_3, and the correlation between log (free lime) and log k are determined statistically. Of the assumptions made, that in which Al_2O_3 was ignored and each molecule of Fe_2O_3 required 0·75 molecules of CaO, was found to give the best result, subject to the limitations involved.

The results of this preliminary investigation indicate that the new technique should be applicable to a wide range of slag-metal problems involved in the steelmaking process.

A Completely Automatic Control of Open-Hearth Reversal. B. M. Larsen and W. E. Shenk. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1830: Preprint, Feb., 1945). A completely automatic system of reversal-valve control for open-hearth furnaces, operated by radiation pyrometers, is described.

The Occurrence of Oxygen in Liquid Open-Hearth Steel—Sampling Methods. T. E. Brower and B. M. Larsen. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1868: Preprint, 1945). The differences in the results of oxygen analyses made from samples of steel taken: (*a*) by different sampling methods, (*b*) at different levels and positions in the bath, and (*c*) from the bath and pouring stream, were studied. Spoon-sampling when carefully carried out is considered preferable to immersing a small mould fitted with a sheet-metal cover in the steel bath by means of a long handle. The oxygen content of liquid steel in a finished heat is substantially the same all over the bath at the same level, but the content tends to decrease at increasing depths.

Electrolytic Manganese in Stainless-Steel Tests at Rustless Steel Corp. and Universal-Cyclops Steel Corp. (United States Bureau of Mines, Oct., 1945, Report of Investigations No. 3829). Tests are described which were carried out at two works making stainless steel rather high in manganese, in order to determine whether electrolytic manganese (about 99·95% purity) could be used in place of low-carbon 80–84% ferromanganese. The electrolytic manganese was produced at the pilot plant of the Bureau of Mines at Boulder City. Both works reported that the use of this material was advantageous in that closer control of the carbon and phosphorus contents was possible whilst the hot-workability, quality, and performance of the steel produced were unaffected.

Electrolytic Manganese in Acid-Steel Tests at Atlas Steel Castings Co. and the Detroit Steel Casting Co. (United States Bureau of Mines, Nov., 1945, Report of Investigations No. 3830). Industrial-scale tests have proved that electrolytic manganese is entirely satisfactory as the form of manganese for making additions to both open-hearth and electric furnace heats of acid steel in the low- and medium-carbon ranges. The proportion of manganese lost was no greater than is the case with ferromanganese.

Production of Nickel and Chrome Steels from Laterite Ores. T. F. Baily. Electrochemical Society, Oct., 1945, Preprint No. 29). Experiments using the electric-arc furnace for the selective reduction of the

laterite nickel-chromium-iron ores of Cuba are described. The principle applied in the experiment was that nickel oxide is more readily reduced than iron oxide, and iron oxide more readily than chromium oxide. Tests showed that only the nickel oxide, part of the iron oxide, and none of the chromium oxide were reduced. The process developed offers commercial prospects where the cost of electric power is low and other favourable conditions obtain.

Composite Steel. (Iron and Steel, 1946, vol. 19, Feb., pp. 57–60). A description is given of the high-frequency electric-furnace plant installed by James Neill and Company (Sheffield), Ltd., for making special steels. There are three furnaces with capacities of 2, 3, and 5 cwt., respectively. Special split moulds with a removable partition down one corner have been devised. High-carbon or alloy steel is first poured into the smaller mould; when this has solidified but is still white hot, the partition is removed, and the remainder of the mould is filled up with soft low-carbon steel. A composite ingot is thus obtained which, when rolled, machined and heat-treated, makes excellent cutting tools.

Stainless Steel Scrap. L. F. Weitzenkorn. (Blast Furnace and Steel Plant, 1945, vol. 33, Dec., pp. 1511–1512, 1537, 1538). A list is given of the American Iron and Steel Institute types and analyses of twenty-nine stainless steels; the value of stainless steel scrap and the necessity for careful segregation of the different types are emphasized.

Scrap-Handling Magnet Improved through Welded Design. (Steel, 1946, vol. 118, Feb. 4, pp. 134–136). Two modifications which improved the efficiency of a magnet for lifting scrap are described. One consisted of closing the gap between the pole ring and the housing, and the other of welding a special central pole shoe to the housing, the shoe being designed to withstand impacts when the magnet is lowered on to the scrap.

Novel Test Sample Speeds Tap Carbon Analysis. (Steel, 1946, vol. 118, Jan. 28, p. 148). A novel casting-pit sampling procedure is described. It consists of taking a test spoonful of slag-free metal from the bath, killing the steel with aluminium wire and drawing a column of approximately 6 in. of clean steel into a 6-mm. pyrex glass tube by suction with a rubber bulb. The sample is quenched in water; the glass is broken away and a sample $\frac{3}{4}$ in. long is sheared from the centre of the rod with clippers.

Progress with Large Electric Steel Furnaces. A. G. Arend. (British Steelmaker, 1946, vol. 12, Jan., pp. 34–37). Advancements in electric-furnace steel-making are reviewed, the subjects dealt with including electro-hydraulic control, making electrode material above the furnace and feeding it continuously through the furnace roof, and rapid charging with baskets.

Steel Production and Its Problems. D. Clark. (Australian Institute of Metals: Australasian Engineer, 1945, Dec. 7, pp. 41–52). A detailed account is given of the production of steel in the arc furnace, the heat-treatment of alloy and high-speed steel, induction hardening, and nitriding. Fatigue, corrosion-fatigue, and the differences between research and investigation are dealt with.

Manufacture of Steel Products from Billet-Size Ingots. R. W. Scholl. (Steel, 1946, vol. 118, Jan. 21, pp. 88–89, 104–108). A description of the steel-making practice at the Connors Steel Co., Birmingham, U.S.A., is given. Steel is made in two electric-arc furnaces with capacities of 3 and $4\frac{1}{2}$ tons/hr. respectively. Small ingots with a maximum size of about $5 \times 5 \times 60$ in. are cast and then rolled into finished products without any necessity for reheating.

Co-Operative Research in the Iron and Steel Industry. C. F. Goodeve. (Sheffield Society of Engineers and Metallurgists: Iron and Coal Trades Review, 1946, vol. 152, Mar. 15, pp. 501–502). The objects and scope of the recently formed British Iron and Steel Research Association are explained. One of its primary objects is to provide a mechanism whereby the brains of the people in the industry could be applied to the benefit of the whole of the industry as well as to that of the particular firm in which they worked. There was, and would continue to be, a Corrosion Testing Station to act as a centre of field investigations in that subject. It was planned to set up a Coatings Research Station at Swansea to specialize in all problems concerned with metallic and other inorganic coatings on steel.

REHEATING FURNACES

Pit Furnaces at Lukens. H. S. Hall. (Iron and Steel Engineer, 1945, vol. 22, Oct., pp. 81–85). An illustrated description is given of an installation of eight soaking-pits at the works of the Lukens Steel Company. Four pits are 15 ft. 9 in. × 10 ft. 6 in., and four are 12 ft. × 10 ft. 6 in., all with a depth of 11ft. 9 in.

Ingot Heating with Modern Soaking Pits. N. W. Richardson. (Iron and Steel Engineer, 1946, vol. 23, Jan., pp. 96–100). The descriptions are given of the following modern types of soaking pits: (1) The one-way-fired recuperative pit; (2) the bottom-fired recuperative pit; (3) the end-fired rectangular recuperative bed; and (4) the circular pit with burners placed tangentially round the bottom.

The Instrumentation of Soaking Pits. J. P. Vollrath. (Blast Furnace and Steel Plant, 1945, vol. 33, Nov., pp. 1391–1393). Descriptions and diagrams are given of electrical and other instruments for measuring and controlling the temperature of soaking pits, the temperature distribution, the pressure, the fuel flow, the gas/air ratio, and for signalling when the soaking process is finished.

Auxiliary Machinery for Ironworks Driven by High-Pressure Oil. M. Oledal. (Teknisk Tidskrift, 1945, vol. 75, Oct. 20, pp. 1139–1144). (In Swedish). Illustrated descriptions are given of charging machines operated by high-pressure oil for putting ingots in reheating furnaces.

FORGING, STAMPING, AND DRAWING

(Continued from pp. 42 A–43 A)

Quality Control of Truck and Bus Forgings. D. A. Barnes. (Steel Processing, 1945, vol. 31, Dec., pp. 760–762). The sequence of processes in the forging of parts of the steering mechanism for heavy motor vehicles in nickel–chromium–molybdenum steel is described and illustrated.

Upsetting Cylinder Barrels. G. W. Birdsall. (Steel, 1945, vol. 117, Dec. 31, pp. 53–56). A description is given of the sequence of processes by which marine engine cylinders about 8 in. in dia. are upset-forged in only four passes from solid blanks with a great saving in machining as compared with former methods.

Photo-Grids. F. Hewlett. (Aircraft Production, 1945, vol. 7, Sept., pp. 425–427). A method of producing fine-lined grids on metal sheets is described. The grids are used for measuring the flow and stretch of metal in pressing and forging operations.

Die Rolling. H. F. Marquardt. (Iron and Steel Engineer, 1946, vol. 23, Jan., pp. 68–71). The manufacture of dies and the process of die-rolling are described in detail.

Forging Die Design. J. Mueller. (Steel Processing, 1945, vol. 31, Dec., pp. 774–776). A description is given of the dies used for forging the crankshaft for a very small petrol engine with a cylinder only $\frac{15}{16}$ in. in dia.

Mass Producing Rocket Projectiles at Westinghouse Plant. W. G. Miller. (Industrial Heating, 1945, vol. 12, Oct., pp. 1694–1700, 1800). The processes of heating, forming, heat-treating, and machining, at a large plant of the Westinghouse Electric Corporation where rocket projectiles are made in huge quantities, are described.

A Review of the German Spring Making Industry. (Wire Industry, 1946, vol. 13, Jan., pp. 33–34). A brief review of the spring industry of Germany is presented; it is based on a Combined Intelligence Objectives Sub-Committee Report by R. W. Cook, supplied by the Controller of Springs of the Ministry of Supply.

New Wire Drawing Machine. (Wire Industry, 1946, vol. 13, Jan., p. 37). A description is given of a new 16-die, dry cone wire-drawing machine for finishing wire of 42–50 S.W.G. in both ferrous and non-ferrous materials.

An Extension of Wire-Drawing Theory, with Special Reference to the Contributions of K. B. Lewis. R. Winstanley Lunt and G. D. S. MacLellan. (Journal of the Institute of Metals, 1946, vol. 13, Feb., pp. 65–96). A rigid analysis of the wire-drawing process through a conical die channel has been developed in a manner formally similar to that adopted by Lewis, but with the introduction of two new quantities. These quantities are: (1) The mean normal stress on the die-channel wall, $\bar{\sigma}_n$, and (2) the mean coefficient of friction, $\bar{\mu}$, between the wire and the die channel, the averaging being carried out with respect to the interfacial surface concerned. The derivation of the numerical values of the two quantities has been described for the case when the wire is fed in tension to the die, and when the reaction on the die seating falls linearly as this tension is increased. By comparing this rigid analysis with Lewis's theory, it is found that his use of $\bar{Y}$, the mean yield stress of the metal in the deformation zone of the die, in place of $\bar{\sigma}_n$, is inadmissible. It is also found that the quantity μ_L, which appears in his theory as the coefficient of friction, is only a numerical factor taking account of the difference between $\bar{Y}$ and $\bar{\sigma}_n$.

The Importance of Reduction of Area in the Evaluation of High Carbon Steel Wire for Cold Drawing Purposes. D. A. Sutch and B. L. McCarthy. (Wire and Wire Products, 1945, vol. 20, Dec., pp. 923–925; 962–963). It is shown that the reduction in area at fracture in the tensile test on patented steel wire is a valuable guide to the microstructure of the material and to its suitability for drawing to make high-quality wire.

Drawing Wire without Lime. H. L. Trembicki. (Wire and Wire Products, 1946, vol. 21, Jan., pp. 38–39). A pickling, washing, and coating process which leaves an excellent lubricant on the wire for subsequent drawing is described. The coating compound has good wetting properties and is of very low alkalinity. Its composition is not stated, but it is a better drawing lubricant than lime.

ROLLING-MILL PRACTICE

(Continued from p. 23 A)

Overload Protection for Mill-Type D.C. Auxiliary Motors. G. A. Moffett. (Iron and Steel Engineer, 1945, vol. 22, Dec., pp. 67–71). The author discusses in detail the operation of several types of overload relays, and their application, particularly to D.C. auxiliary motors in the rolling mill.

Cold Rolling. C. E. Davies. (Metallurgia, 1946, vol. 33, Jan., pp. 147–153). Recent progress in the design of rolling mills and in the cold-rolling practice of steel strip and sheets in Great Britain are reviewed.

Roll Manufacture and Application. A. E. Murton. (Iron and Steel Engineer, 1945, vol. 22, Nov., pp. 51–56). The author describes the types of rolls in common use in American rolling mills and briefly discusses their manufacture, and the rolling conditions for which each type is suitable.

Merchant Mill and Alloy Finishing Facilities at Fontana. R. M. Bickerstaff. (Blast Furnace and Steel Plant, 1945, vol. 33, Dec., pp. 1513–1522). A detailed description is given of the merchant bar mill

at the Kaiser Steel Plant, Fontana, California. It was put in commission in April, 1945, and is designed to roll flats up to 18 in. wide, rounds up to 4½ in. in dia., squares up to 3 × 3 in., and angles up to 3 × 3 in. The mill has a capacity of about 180,000 tons per annum.

The Manufacture of Seamless Steel Tubes. J. W. Jenkin. (Sheffield Society of Engineers and Metallurgists: Metallurgia, 1945, vol. 33, Nov., pp. 3–6; Dec., pp. 59–64; 1946, Jan., pp. 139–144). Detailed and illustrated descriptions are given of various methods of manufacturing seamless steel tubes including the pilger, Diescher, Assel and Foren processes.

Seamless Steel Tubes. H. Sanders. (British Steelmaker, 1945, vol. 11, Nov., pp. 500–504; Dec., pp. 557–561). Descriptions are given of the processes involved in making "hollows" (the raw material for seamless steel tubes), the sinking and plug-drawing methods of reducing the hollows and the heat-treatment of the finished tubes.

Heavy Wall Pipe. E. C. Wright. (American Society for Testing Materials: Steel, 1945, vol. 117, Dec. 31, pp. 70–72). The difficulties and high costs involved in the rolling of thick-walled tubes which are only required in small quantities are discussed, and some suggestions for simplifying the specifications and reducing the costs are made.

Salvaging Forged Alloy Steel Rolls. H. L. Watson, jun. (Iron and Steel Engineer, 1945, vol. 22, Dec., pp. 55–60). The following methods of salvaging worn-out rolls are described and illustrated: Simple regrinding after normal use; repairing damaged rolls by grinding and etching; stress-relieving treatments; rehardening for use in the same mill; turning down into smaller rolls for use in another mill; renewing either forged or cast back-up rolls by shrinking-on a sleeve.

HEAT-TREATMENT

(Continued from pp. 43 A–45 A)

Heat Treating B-29 Super-Fortress Engine Parts in Mammoth Dodge-Chicago Plant. (Industrial Heating, 1945, vol. 12, Oct., pp. 1658–1690; Nov., pp. 1872–1878, 1900). An illustrated description is given of an extensive heat-treatment plant of the Chrysler Corporation. This comprises carburizing furnaces, a continuous gas-fired radiant-tube annealing furnace 60 ft. long, a nitriding plant, and continuous hardening and tempering furnaces. The heat-treatment plant for making and maintaining the tools for the above operations are described in the second part of the article.

Fabricating and Heat Treating Hollow Steel Propeller Blades at Curtiss-Wright Plant. (Industrial Heating, 1945, vol. 12, Oct., pp. 1702–1716; Nov., pp. 1888–1900). The equipment and processes used in the manufacture of hollow steel propeller blades for aircraft are described. This involves milling and blanking the plates with subsequent pressing, shaping, welding, brazing, and heat-treatment.

Grinding Hardened Steel Precision Gears. E. A. Koether. (Engineering Materials and Processing Methods, 1945, vol. 22, Dec., pp. 1754–1758). The heat-treatment and finishing of gears for aero-engines are discussed with special reference to the effects of grinding, decarburization and shot-peening on the fatigue strength.

Temperature-Time Distribution in Rectangular Bars. W. M. Rohsenow, M. J. Aronstein, and A. C. Frank. (Transactions of the A.S.M.E., 1946, vol. 68, Feb., pp. 135–141). Charts for determining the changes of temperature with time in the heating of flat plates, infinite cylinders, and spheres have been published previously. In this paper the lower ends of these curves are amplified. In addition, the rapid convergence of the series solution is shown, and series of curves for rectangular bars are presented.

Gaseous Cementation. J. Pomey. (Laboratoire des Usines Renault, Billancourt, Paris, Dec. 18, 1942). A comprehensive report on the gas-carburization of steel is presented in four parts. In Part I. the history and present state of development of the process are reviewed and the equilibria between iron, austenite and cementite on the one hand, and CO-CO_2 mixtures and the hydrocarbon gases on the other hand, are dealt with. In Part II. an account of laboratory experiments with gas-carburizing is given. In Part III. the preparation of a suitable carburizing gas and the control of its temperature, pressure and composition by methods developed at the Usines Renault are described; this gas consists of charcoal producer gas with an addition of propane. In Part IV. a small rotating chamber for carburizing small parts in batches and a large continuous furnace equipped with a pusher for dealing with plates are described.

Summary of a Study of Gaseous Cementation. J. Pomey. (Laboratoire des Usines Renault, Billancourt, Paris). The equilibria involved in gas-carburizing are studied and the plant developed at the Usines Renault for preparing the gas and carburizing small parts and plates is described. This process has been found to cost less than half as much as box-carburizing with a solid medium.

Characteristics and Uses of Salt Baths. R. C. Stewart. (Iron Age, 1946, vol. 157, Feb. 14, pp. 46–52). The development of salt baths with submerged water-cooled electrodes has given new impetus to the use of salt baths in the mass-production heat-treatment of both ferrous and non-ferrous metals. In this paper a review is presented of these and other improvements in cyanide and carburizing baths, with notes on the composition of the solutions used.

The Isothermal Heat-Treatment of Steel. E. F. Watson and J. McArthur. (Machinery, 1946, vol. 68, Feb. 28, pp. 265–271). The isothermal heat-treatment of steel is explained by reference to the S-curve. The equipment required and the properties of the molten salt for quenching are discussed, and the

results obtained by this treatment in the hardening of rocket tubes, knife blades, machete blades, and pistol parts are given. The principal advantages of the treatment are the reduction in distortion and the fact that finished-machined parts can be hardened.

Precise Control of Steel Hardening. H. A. Knight. (Engineering Materials and Processing Methods, 1945, vol. 22, Dec., pp. 1736–1738). A brief description is given of the heat-treatment furnaces, the process, and the methods of control which were developed for hardening 24½-in. bars, tapering from ¾ in. to 1¼ in. in dia., of chromium–molybdenum steel.

Effect of Preliminary Treatment on High-Speed Steels. F. F. Dodson. (Metallurgia, 1946, vol. 33, Jan., p. 124). The sodium-cyanide treatment for high-speed steel described by J. McIntyre (*see* Journ. I. and S.I., 1945, No. I., p. 50 A) is discussed and some experiments on 18%-tungsten and 6%-molybdenum high-speed steels are described. It was confirmed that the treatment does slightly enhance the hardenability for a certain depth, but does not increase the total hardenability.

Flame-Hardening in the Small Shop. W. B. Sharav. (Welding Journal, 1946, vol. 25, Jan., pp. 19–24). A number of designs of flame-hardening equipment for use in small workshops are described and illustrated.

Flame-Hardening Small Machined Parts. (Iron Age, 1946, vol. 118, Feb. 7, p. 77). A flame-hardening machine for parts between 1½ and 12 in. in dia. is described. The part to be hardened is spun on a vertical axis and heated by a number of radially positioned oxy-acetylene burners. On attaining the correct temperature the burners are withdrawn and sprays of water are directed on the part.

Induction Hardening as Applied to Farm Machinery. F. F. Vaughn. (S.A.E. Journal, 1946, vol. 54, Jan., pp. 38–43). Problems involved in induction-hardening are reviewed and some examples of the successful application of induction-hardening to parts of agricultural machinery are described.

Induction Heating. N. R. Stansel. (Iron and Steel Engineer, 1946, vol. 23, Jan., pp. 102–111). The principles of induction heating are explained and data which enable the heating rates of different sizes and shapes of bar to be calculated are presented.

Induction Heating for Shrinking-On Heavy Parts. (Machine Shop Magazine, 1946, vol. 7, Feb., pp. 45–46). Descriptions are given of two induction-heating machines made up from components of resistance-welding machines. The former were used for rapidly heating up 50-lb. steel suspension arms for shrinking on to spindles.

Nitriding of Aircraft Steels. P. A. Haythorne. (Iron Age, 1946, vol. 157, Jan. 31, pp. 44–48). An investigation is described in which the surface hardness and changes of hardness with depth obtained on nitriding the following steels at 900°, 950°, and 1000° F. in 30%, 45%, and 60% dissociated ammonia, respectively, were studied : (1) A 0·19%-carbon steel, (2) a 0·95%-carbon steel, (3) a 0·91%-chromium 0·22%-molybdenum steel, (4) three chromium–nickel–molybdenum steels, and (5) Nitralloy G (containing 1·33% of chromium, 0·33% of molybdenum, and 1·30% of aluminium). The results are presented as depth-hardness curves. The acceptance of nitrided steels on the basis of surface hardness alone is not generally justified, although this value may serve as a basis for comparison.

Cycle Annealing of Hypo-Eutectoid Steels. G. R. Brophy. (Iron Age, 1945, vol. 156, Dec. 13, pp. 69–71). The austenite in some complex alloys is so slow to transform during continuous cooling that many days are required to soften them adequately for machining and cold-working. A method of heat-treatment is outlined which reduces the time of treatment.

The Application of Cathode-Ray Oscillographs when Photographing Cooling Curves for Quenching Oils. B. Levy. (A/B. Wahlén och Block, Wabo Tekniska Meddelanden, 1945, No. 3, pp. 1–6). (In Swedish). A description is given of an apparatus which was set up for obtaining permanent records of the cooling curves for quenching oils obtained with a cathode-ray oscillograph.

The Resistance of Quenching Oils to Oxidation. (A/B. Wahlén och Block, Wabo Tekniska Meddelanden, 1945, No. 3, pp. 6–8). (In Swedish). A description is given of a simple appartaus for determining the rate at which quenching oils absorb oxygen. The stability of an oil is greatly reduced by an organic compound, ferric naphthenate, which forms after the oil has been used a few times. Rust also reduces the resistance to oxidation, but to a less extent. The use of oxidation inhibitors is discussed.

WELDING AND CUTTING

(Continued from pp. 45 A–47 A)

Economics of Arc Welding. W. J. Brooking. (Steel, 1946, vol. 118, Jan. 21, pp. 81, 111, 112 ; Jan. 28, pp. 124–126, 187–189). The costs of electric welding are analysed and several important factors affecting the costs are discussed.

Arcwelding Trends in Wartime Germany. F. W. Myers, jun. (Iron Age, 1946, vol. 118, Feb. 7, pp. 56–60). The measures which were taken in Germany during the war to meet the increased demands on the welding industry are outlined. No very important developments are cited, but some interesting work was done in the production of welded passenger coaches, frameless tank cars, and mine tubs.

Metallurgy as Applied to Welding. R. B. Lincoln. (Welding Journal, 1945, vol. 24, Dec., pp. 1143–1147). The metallurgical characteristics of carbon steels and alloy steels which affect its welding properties are explained. If electrodes of 14%-manganese steel are used to deposit a hard face on plain carbon steel there will be a layer, somewhere between the parent metal and the surface, where the composition is 3–5% of manganese and this will be hard and brittle. This

can be avoided by depositing an intermediate layer of 18/8 stainless steel.

Metallurgical Features in Welded Steels. H. O'Neill. (Transactions of the Institute of Welding, 1946, vol. 9, Feb., pp. 3–9). The conditions affecting the weldability of high-tensile steels are reviewed. A progress report is given on the new simple weld ductility test. From a study of the welding of steel both in the primitive smith's hearth and by modern non-fusion processes a critical definition of the art is suggested. This definition would distinguish in principle between bonding effects due to true welding and adhesion effects such as those which occur between loaded metallic surfaces in rubbing contact. The fusion welding of ordinary steels is considered next with particular reference to the curious local hardening produced in ferrite by "stray flashes". Wöhler fatigue tests on a low-alloy steel intentionally provided with these hard spots have indicated the danger of such isolated arc craters in structures subjected to dynamic stresses.

Selection of Steel for Welding. S. L. Hoyt. (Welding Journal, 1945, vol. 24, Dec., pp. 1180–1187). Factors affecting the selection of steel for welding are discussed with special reference to steels for oil-well casings and shipbuilding.

Arc Welding at the Newcastle Steel Works. A. Young. (B. H. P. Review, 1945, vol. 23, Dec., pp. 18–19). Examples of the electric welding for construction and maintenance at the Newcastle Steel Works of Broken Hill Proprietary Co., Ltd., are described and illustrated.

The Quantitative Measurement of Welding Response by Bead Welds. R. D. Stout, S. S. Tör, and G. E. Doan. (Welding Journal, 1945, vol. 24, Dec., pp. 625-S–628-S). In the system for predicting the ductility of welds developed at Lehigh University, the Jominy end-quench test was used to measure the response of the steel to welding. This is an indirect method of measurement. In this paper a direct method is proposed. A series of welds are made by depositing a bead on each of a number of plates, using a different rate of heat input each time. Hardness traverses are made on each. By applying a "geometry factor" to the hardnesses measured for the various rates of heat input it is then possible to predict the heating rates which would produce the desired hardness in a weld of different design.

The Welding of Nickel-Alloy Steels. T. N. Armstrong. (Iron Age, 1946, vol. 157, Jan. 24, pp. 36–41; Jan. 31, pp. 49–55; Feb. 7, pp. 69–76; Feb. 14, pp. 56–60). In the first part of this series of articles the factors influencing the weldability and tests for the weldability of nickel steels are dealt with. In the second part weldability tests, including the T bend test, the single V-groove weld test, and the Reeve cracking test, and tests for selecting welding electrodes are described. In the third part structural welding and the welding of pressure vessels are dealt with. In the fourth part descriptions are given of spot-welding, upset butt-welding, forge welding, oxy-acetylene welding, and oxy-acetylene pressure welding.

Welding of Nitralloy Steels. D. Llewellyn. (Welding, 1946, vol. 14, Feb., pp. 70–73). The results obtained from hardness, tensile and impact tests on welded specimens of low-alloy chromium–nickel–molybdenum steels to British Standard Specifications En 40 and En 41 are presented and different methods of welding these Nitralloy steels are discussed. The atomic hydrogen welding process is the most satisfactory; oxy-acetylene welding gives four results, and metallic-arc welding is satisfactory provided that suitable electrodes are used.

Resistance Welding on Stainless Steel Aircraft Engines and Accessories. F. G. Harkins and W. L. Hales. (Welding Journal, 1946, vol. 25, Jan., pp. 39–46). Descriptions are given of methods adopted for the resistance-welding of complicated stainless-steel parts for aircraft engines.

Resistance Welding. Part I. Modern Developments and Applications in Up-to-Date Fabrication Methods. Part II. Equipment for Spot Welding Light Alloy Material: Stitch and Projection Welding. Part III. A Survey of the Seam Welding Process and Equipment. R. W. Ayers. (Aircraft Production, 1945, vol. 7, Aug., pp. 379–384; Sept., pp. 446–450; Oct., pp. 490–494). A comprehensive and profusely illustrated review of the latest types of spot- and seam-welding machines for both ferrous and non-ferrous metals is presented and the principles employed are explained.

Measurement of the Effect of Contact Resistance in Spot Welding. R. A. Wyant. (Electrical Engineering, 1946, vol. 65, Jan., pp. 26–33). A description is given of an apparatus for measuring the electrical contact resistance in the spot-welding of sheet metal. The results of resistance tests in the spot-welding of aluminium-clad steel sheet are presented and discussed.

Quality Control and Spot Welding. N. G. Dod. (Aircraft Production, 1945, vol. 7, Dec., pp. 563–564). The application and value of statistical methods in controlling the quality of spot welds are discussed.

Testing Welds in Holland. (Institute of Welding: Metallurgia, 1946, vol. 33, Jan., p. 145). A brief account is given of a co-operative organization which the shipbuilding and engineering industries of Holland have set up for making X-ray examinations of welded structures of all kinds, particularly for shipbuilding.

Controlling Distortion in the Gas Welding of Sheet Metals. C. G. Bainbridge. (Sheet Metal Industries, 1946, vol. 23, Feb., pp. 356–362). The causes of distortion in the gas-welding of sheet metal are analysed. The correct preparation of butt, corner, lap and T joints and the appropriate welding procedure are described, and details of jigs for particular jobs are given.

How to Select and Use Cast Iron Shanks for Cutting Tools. (Steel, 1946, vol. 118, Jan. 14, pp. 90–91, 124–126). A description is given of the making of cutting tools by brazing high-speed-steel or tungsten-carbide tips to shanks of high-strength cast iron.

Modern Hard-Facing Methods. A. R. Lytle. (Welding Journal, 1945, vol. 24, Dec., pp. 1155–1165). Illustrated descriptions are given of the equipment and techniques for welding a layer of abrasive and/or heat-resistant alloys on to the wearing surface of steel bars, cylinders, seating rings, and gears.

Unusual Flame Cutting. (Welding, 1946, vol. 14, Jan., pp. 40–42). Descriptions are given of the oxy-

acetylene cutting of nickel–chromium–molybdenum steel 44 in. and 54 in. thick. A test was also made on mild steel 60 in. thick which demonstrated that it was a practical proposition with the ½-in. dia. nozzle employed. Details of the cuts, the equipment used, and the oxygen consumption are given.

Flame Cutting. J. F. Smythe. (Steel, 1946, vol. 118, Jan. 14, pp. 104, 131). An 8-torch electrically controlled flame-cutting machine was developed for the mass production of breech blocks. A description of the machine is given.

Cost Reduction Aspects of Flame-Cutting Operations. G. V. Slottman. (Machinist, 1946, vol. 89, Feb. 16, pp. 2392–2395). Methods of increasing production and decreasing costs when using multiple-burner oxy-acetylene cutting machines are described and illustrated.

PROPERTIES AND TESTS

(Continued from pp. 50 A–54 A)

The Effect of Triaxiality on the Technical Cohesive Strength of Steels. G. Sachs and J. D. Lubahn. (American Society of Mechanical Engineers: Welding Journal, 1946, vol. 25, Jan., pp. 53-S–63-S). The fundamental characteristics of metals revealed by means of extensive tensile tests on notched bars are described. The effects of the three major factors, triaxiality, plastic strain, and stress concentration, can be separated by means of systematic tests covering the dependence of the strength of notched bars upon such variables as notch radius, notch depth, and hardness. The characteristic strength values obtained by tensile tests on notched bars appear to be of considerably greater practical significance than those obtainable by means of the conventional tensile test.

Cold Heading Die Steel. A. S. Jameson. (Steel, 1946, vol. 118, Jan. 21, pp. 82–85, 123–126). Data on the hardness, hardness depth, and grain size of high-carbon and low-alloy steels for cold-heading dies are presented and discussed.

The Fabrication of High-Strength Steel. C. E. Loos. Steel Processing, 1945, vol. 31, Dec., pp. 755–759). The hot- and cold-working properties of two high-strength steels, namely, " Cor-Ten " and a structural carbon steel are compared.

The Effect of Various Elements on the Hot-Workability of Steel. H. K. Ihrig. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1932: Metals Technology 1945, vol. 12, Oct.). A quantitative hot-workability test has been devised to determine the effect of various elements on the hot-workability of steels. The following elements had little or no effect: Oxygen, carbon, phosphorus, cobalt, vanadium, and titanium. The following elements had a beneficial effect: Manganese, nickel, and chromium (above 9%). The following elements had a detrimental effect: Sulphur, selenium, silicon, nitrogen, molybdenum, columbium, lead, tin, and chromium (below 9%).

Effect of Heat Treatment on the Endurance Limit of Alloyed Gray Cast Iron. T. E. Eagan. (American Foundryman, 1945, vol. 8, Dec., pp. 44–53). Tests were carried out to determine whether nickel–molybdenum cast iron could be heat-treated so as to obtain sufficiently high endurance limits for the material to be used for making crankshafts. The results indicated that, in general, heat-treatment to obtain a high tensile strength does not increase the endurance limit in proportion. The original as-cast structure has a profound influence on the endurance limit obtained after heat-treatment. Nickel–molybdenum and nickel–chromium–molybdenum grey cast irons with not much free cementite present can be heat-treated to give an acicular structure with an increase in the endurance limit.

Addition Method of Calculating Rockwell C Hardness of the Jominy Hardenability Test. W. Crafts and J. L. Lamont. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1928: Metals Technology 1945, vol. 12, Oct.). A method of calculating the hardness of alloy steel in Rockwell C units is described. A basic hardness number corresponding to that of a pure iron-carbon alloy cooled at a certain rate is taken, and to this increments of hardness, corresponding to the amounts of alloying elements present, are added. The validity of the calculation of the hardness was confirmed by data obtained on standard qualities of steel and the accuracy is sufficient for the method to be useful in predicting and controlling hardenability.

An Appraisal of the Factor Method for Calculating the Hardenability of Steel from Composition. G. R. Brophy and A. G. Miller. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1933: Metals Technology 1946, vol. 12, Oct.). Grossmann's method of calculating the hardenability of steel from its composition is critically reviewed. A number of tests are described in which the effect of nickel, manganese, chromium and molybdenum on the hardenability of steel were determined, and the data obtained are compared with the hardenabilities predicted by the above method.

Supersonic Testing in a Steel Plant. J. V. Russell and H. E. Pellett. (Iron Age, 1946, vol. 157, Jan. 31, pp. 38–43). A description is given of the Sperry supersonic reflectoscope for detecting flaws in billets and blooms up to 20 in. square. The presence and position of defects are determined by measuring the difference in the times at which reflections of waves of supersonic frequency transmitted from one side of the billet are received from the opposite surface and from the defect.

The Detection of Cracks in Steel by Means of Supersonic Waves. C. H. Desch, D. O. Sproule, and W. J. Dawson. (Iron and Steel Institute, 1946, this Journal, Section I.). In Part I. previous methods of detecting flaws in steel by means of sound waves, especially those of high frequency, are reviewed. In Part II. a description is given of an apparatus employing piezo-electric quartz transmitters for pro-

ducing intermittent pulses of supersonic waves, and receiving them on similar quartz crystals after reflection, the reflected waves being electrically amplified and indicated on a cathode-ray oscilloscope. The conditions for obtaining sharp indications are discussed, as well as the effects of the degree of finish of the metallic surfaces under examination. In Part III. are given the results of applying the new method in works practice, including those obtained with large masses, billets, plates, castings, and welds. The conditions necessary for the satisfactory use of the apparatus for practical testing are discussed. It is maintained that supersonic testing should be regarded primarily as a new instrument of investigation, giving information which cannot at present be obtained by more usual methods.

Supersonic Inspection. (Steel, 1945, vol. 117, Dec. 24, pp. 94–96). A description is given of the " Hypersonic Analyser " for the detection of minute defects in steel. The instrument is capable of transmitting wave frequencies in the 50–1000 kilocycle range.

Magnetic Analysis Inspection of Metals. H. L. Edsall. (Engineering Materials and Processing Methods, 1945, vol. 22, Dec., pp. 1731–1735). Examples are given of the application of electronics to detect defects in steel bars and tubes by irregularities in their magnetic properties.

Sorting Metals by Tribo-Electrification. A. Doscheck. (Steel, 1945, vol. 117, Dec. 24, pp. 106–110). An electrical method of sorting metals is described. It is based on the measurement of the potential produced when the specimen of unknown origin is made to vibrate against a reference specimen. If the specimens are identical no potential is produced, but if they are different, the millivoltage and the polarity are indicative of the differences in composition or structure.

Fracture Phenomena in Steels at Elevated Temperatures. C. Schaub. (Jernkontorets Annaler, 1946, vol. 130, No. 1, pp. 1–26). (In Swedish). After reviewing several recent papers on creep tests, particularly on the types of fracture obtained in creep tests, an account is given of investigations carried out in 1942–1944 at the Kungl. Tekn. Högskolans institution för hållfasthetslära in Stockholm. In order to obtain a complete picture of the behaviour of steel at high temperature the usual creep-test should be supplemented by impact tests after long-time annealing and by long-time tensile tests on notched specimens. A stable material should have no embrittlement tendency and there should be no tendency to form cracks similar to stress-corrosion cracks. Tests with a 0·7%-chromium 3·25%-nickel 0·65%-molybdenum steel showed that, by raising the chromium content at the surface to about 30%, the life of the steel at high temperatures was increased and there was no tendency to crack formation.

Graphitization of Carbon–Molybdenum Steel in High-Temperature Steam Piping. G. A. Timmons. (American Society of Mechanical Engineers: Welding Journal, 1946, vol. 25, Jan., pp. 1-S–6-S). Graphitization of carbon-molybdenum steel tubes in high-temperature steam installations is considered from the viewpoints of thermodynamic stability and rates of reaction. Cementite is not stable below 1490° F. but graphite is, so that graphitization may be expected whenever cementite is present in a steel placed in service at 850–1000° F. Factors which accelerate graphite nucleation lengthen service life. By restricting the carbon content to less than 0·06% and providing alloying elements to give adequate creep strength it should be possible to produce steel suitable for high-temperature steel tubes. A complex carbide $(FeMo)_{23}C_6$ may be produced in a low-carbon 0·50%-molybdenum steel by the isothermal transformation of austenite at 1300° F. Some creep-test data for iron alloys containing 0·50% of molybdenum and less than 0·05% of carbon are presented.

Opinions on Graphitization. L. Crome. (Foundry, 1945, vol. 73, Dec., pp. 102–103, 248–258). The mechanism of the formation and growth of graphite in cast iron and of the solution of graphite in white iron is explained and discussed.

METALLOGRAPHY AND CONSTITUTION

(Continued from pp. 54 A–55 A)

The Upright Type of Microscope in Metallography. J. R. Vilella. (Metal Progress, 1946, vol. 49, Jan., pp. 113–116). The merits of the inverted and upright types of microscope are discussed and micrographs of the same area of metallographic specimens taken with both types of instrument are compared.

A New Etching Reagent for the Detection of Oxygen Segregations in Steel. L. Fine. (Metal Progress, 1946, vol. 49, Jan., pp. 108–112). Details are given of the preparation of an alkaline chromate etching solution and examples are cited illustrating its successful use to detect and study segregated oxides in steel.

Grain Boundaries. R. C. Bernhoeft. (Metallurgia, 1946, vol. 33, Jan., pp. 117–120). The theories relating to grains and grain boundaries in the structure of metals, which have appeared in the literature since 1929 are reviewed and discussed.

Solid Solutions—Classification of the Solubilities of Elements in Iron. I. I. Kornilov. (Iron and Steel, 1946, vol. 19, Feb., pp. 52–56). This is a translation of a Russian paper which appeared in Izvestia Akademii Nauk S.S.S.R. The author studies the solubilities of elements in iron in the solid state from the point of view of Hume-Rothery's size-factor theory. He reviews the available information and shows that the size factor alone is adequate to account for the experimental results except in the case of cerium and certain elements of the B sub-groups of the fourth and fifth periods of the periodic system.

The Phase Changes of SAE 2345 Steel. A. Dubé and R. L. Cunningham. (Transactions of the Canadian Institute of Mining and Metallurgy, 1946, vol. 49, pp. 1–16). The mechanical properties of steel S.A.E. 2345 (a 3·5%-nickel steel) after quenching and

tempering at 1050°, 1150°, 1200°, and 1225° F. were studied. A hardening and embrittling effect on the steel was noted when the steel was tempered at temperatures higher than 1150° F. On holding the steel at temperatures somewhat below the Ac_1 point (1275° F.), as determined with a dilatometer, there was a decrease in the intensity of magnetization as well as in specific volume. X-ray diffraction technique revealed that the phase formed on holding at 1240° F. was austenite, and this austenite had a sluggish transformation behaviour.

CORROSION OF IRON AND STEEL

(Continued from pp. 55 A–56 A)

Developments in Corrosion Studies and Corrosion Control. H. M. Olson. (Iron and Steel Engineer, 1946, vol. 23, Jan., pp. 80–95). A review of the literature from 1935 to 1945 on corrosion studies is presented.

Some Notes on Corrosion in Cast-Iron Sectional Boilers. E. R. Walter. (Journal of the Institution of Heating and Ventilating Engineers, 1946, vol. 13, Jan.–Feb., pp. 285–292). The causes of the corrosion of low-pressure cast-iron sectional boilers are examined and suggestions for its prevention are made. To prevent the steam in the products of combustion from condensing, all parts of the heating surface must be maintained at a temperature above the dew-point. The return-water temperature should be raised by by-passing a portion of the hot outlet water into the return manifold.

Stress Corrosion Cracking of Mild Steel. Part II. The Accelerated-Precipitation Theory of Stress Corrosion. J. T. Waber and H. J. McDonald. (Corrosion and Material Protection, 1945, vol. 2, Dec., pp. 13–16). A general theory of stress corrosion has been developed. Contrary to previous conceptions the present theory states that the direction of crack growth depends upon the localized electrochemical condition which has been induced by the applied and residual forces.

Stress Corrosion Cracking of Mild Steel. Part III. Methods of Testing the Accelerated-Precipitation Theory. J. T. Waber, H. J. McDonald and B. Longtin. (Corrosion and Material Protection, 1946, vol. 3, Jan., pp. 13–19). The conditions which an alloy must satisfy in order to be susceptible to stress corrosion are set forth. These include that the alloy must be capable of strain-ageing and that one of the precipitated phases is anodic to the parent phase, the latter not being severely attacked by the corrosive environment. It is shown from thermodynamic considerations that Fe_4N is anodic to steel and can form $Fe(OH)_2$ rather than Fe_3O_4.

Reaction Rate of Hydrogen Chloride and Sulfide with Steel. C. F. Prutton, D. Turnbull, and G. Dlouhy. (Industrial and Engineering Chemistry, Industrial Edition, 1945, vol. 37, Nov., pp. 1092–1097). In investigations of the possible corrosive action of extreme pressure lubricants on gear teeth the rate of attack on iron and steel by hydrogen chloride, hydrogen sulphide, and mixtures of the two were carried out. At a pressure of 1 atm. the amount of reaction, y, of iron with hydrogen sulphide may be represented as a function of time, t, by the equation $y^2 = Kt + A$, where K is a constant depending upon temperature and pressure, and A is a constant which may be zero or have a positive value depending on the type of iron. Hydrogen chloride reacts much less rapidly with iron and steel than hydrogen sulphide. When a 1 to 1 mixture of the two reacts with steel, the rate of formation of iron chloride is about the same as when hydrogen sulphide is absent.

Total-Immersion Apparatus. M. Cohen. (Electrochemical Society: Steel, 1946, vol. 118, Jan. 14, p. 106; Iron and Steel, 1946, vol. 19, Feb., pp. 75–76). A description is given of an apparatus for making total-immersion corrosion tests in which the specimen is held in a glass tube connected by rubber tubing to two graduated flasks. One flask is kept stationary and the other is moved up and down by a motor-driven mechanism, thus causing the water, or other corroding medium, to flow backwards and forwards over the surface of the specimen.

ANALYSIS

(Continued from pp. 33 A–34 A)

Iron and Steel Analysis. E. C. Pigott. (Iron and Steel, 1945, vol. 18, Dec., pp. 641–645; 1946, vol. 19, Jan., pp. 7–12; Feb., pp. 61–62). Details are given of methods recommended by the author for determining the following elements in iron and steel: Aluminium, carbon, chromium, cobalt, copper, lead, manganese, molybdenum, nickel, phosphorus, silicon, sulphur and zirconium.

A Photoelectric Recording Polarograph. L. Lykken, D. J. Pompeo, and J. R. Weaver. (Industrial and Engineering Chemistry, Analytical Edition, 1945, vol. 17, Nov., pp. 724–728). The apparatus described is a convenient photo-electric recording polarograph for accurate analytical work; the polarograms are automatically traced on a moving chart in plain view of the operator.

Photoelectric Type Portable Photometers. (British Standards Institution, 1945, No. 667). This specification covers the requirements for portable photoelectric photometers in which the test surface embodies a rectifier photo-electric cell.

Evaluation of Spectrographic Analytical Data. H. V. Churchill and J. R. Churchill. (Industrial and Engineering Chemistry, Analytical Edition, 1945, vol. 17, Dec., pp. 751–754). The statistical treatment of data obtained by spectrographic analyses is

discussed. The magnitude of the standard deviations is in general no higher than would be expected had the comparisons been made among chemical analyses carried out by different routine analysts.

Semi-automatic Spectrographic Analysis. R. H. Zachariason. (Industrial and Engineering Chemistry, Analytical Edition, 1945, vol. 17, Dec., pp. 749–750). Semi-automatic devices are described which simplify spectrographic analyses by reducing considerably the time required for changing the plate after each exposure, focusing the electrodes, controlling the exposure, and changing the electrodes.

Spectrographic Analysis. J. Convey and L. G. Young. (Journal of the Birmingham Metallurgical Society, 1945, vol. 25, Sept., pp. 168–189). Current literature indicates that instability of the source of light and insufficient reproducibility of spectral excitation used by different experimenters are the main difficulties which to-day retard the development of applied spectrography. In this paper spectrographic excitation sources are discussed and methods of stabilizing spectral emission are described.

Spectrochemical Determination of Small Amounts of Aluminium in Steel. R. Spiers, P. Fischer, and K. L. Proctor. (Industrial and Engineering Chemistry, Analytical Edition, 1945, vol. 17, Dec., pp. 772–774). A spectrochemical method is described for the determination of aluminium in steel samples weighing from 0·020 to 0·500 g. Colorimetric procedures proved to be unsuitable because of the presence of interfering ions. In the method developed, a mercury cathode separation is employed and one drop of the aluminium solution, prepared by the chemical procedure described, is placed on each of the flat-topped carbon electrodes which are sparked for 75 sec. immediately after the solution has been absorbed.

Spectrophotometric Determination of Titanium in Steels. A. Weissler. (Industrial and Engineering Chemistry, Analytical Edition, 1945, vol. 17, Dec., pp. 775–777). Absorption spectra have been determined for the hydrogen-peroxide complexes of titanium and vanadium in the presence of 1 g. of iron, in perchloric-phosphoric acid solution. These spectra furnish the basis for improved rapid and accurate methods for the determination of from 0·5 to 6 mg. of titanium in iron, steel, vanadium steel, and stainless steel, without separating the titanium or using hydrofluoric acid.

Rapid Method for Determination of Copper in Steel. F. B. Clardy, J. C. Edwards, and J. L. Leavitt. (Industrial and Engineering Chemistry, Analytical Edition, 1945, vol. 17, Dec., pp. 791–792). A method is described for the rapid and accurate determination of copper in plain and alloy steels for foundry control. The total time required is approximately 15 min.

The Spectrochemical Determination of Metallic Impurities in Steel. C. G. Carlsson. (Iron and Steel Institute, 1946, Translation Series, No. 267). An English translation is presented of a paper which appeared in Jernkontorets Annaler, 1945, vol. 129, No. 5, pp. 193–241 (*see* Journ. I. and S.I., 1945, No. II., p. 151 A).

Polarographic Determination of Nickel in Steel and Nickel Ore. P. W. West and J. F. Dean. (Industrial and Engineering Chemistry, Analytical Edition, 1945, vol. 17, Nov., pp. 686–688). A rapid and accurate method for determining nickel in steel and in nickel ores, based on the use of the polarograph with sodium fluoride as the supporting electrolyte, is described.

Determination of Nitrogen by Combustion. H. Gonick, D. D. Tunnicliff, E. D. Peters, L. Lykken, and V. Zahn. (Industrial and Engineering Chemistry, Analytical Edition, 1945, vol. 17, Nov., pp. 677–682). Improved apparatus and procedure are described for determining nitrogen by Dumas' combustion method.

BOOK NOTICES

(Continued from p. 56 A)

HARVEY, H. W. "*Recent Advances in the Chemistry and Biology of Sea Water.*" 8vo, pp. viii + 164. Cambridge, 1945: University Press. (Price 10*s.* 6*d.*)

A summary of the large amount of important work that has been done in recent years on the relation of plant and animal life in the sea to the chemical constituents of sea water. Much of this work has been carried out by Dr. Harvey himself and other members of the staff of the Marine Biological Laboratory at Plymouth.

MANTELL, C. L. and C. HARDY. "*Calcium Metallurgy and Technology.*" 8vo., pp. 145. Illustrated. New York, 1945: Reinhold Publishing Corporation. (Price 20*s.*)

A survey of available information on the properties, production and uses of calcium, and the constitution of the alloy systems. A chapter is devoted to the use of calcium in ferrous metallurgy.

PULL, E. "*Boiler House Practice.*" Second Edition. Sm. 8vo., pp. vi + 282. London, 1945: The Technical Press Ltd. (Price 10*s.* 6*d.*)

This book is intended as a guide to the selection, maintenance and operation of industrial steam generators and their auxiliary equipment. It also deals in a practical manner with problems relating to the efficiency of such plants.

TIBBENHAM, L. "*The Welding of Cast Iron by the Oxy-Acetylene Process.*" 2nd Edition. Sm. 8vo., pp. viii + 120. Illustrated. London, 1945: Sir Isaac Pitman and Sons Ltd. (Price 7*s.* 6*d.*)

This book deals in a practical way with every detail of the process of the oxy-acetylene welding of cast iron. Such subjects as gases employed, varieties of generator plants, composition of welding rods and fluxes, vital facts in the metallurgy of cast iron relating to welding, and methods of pre-heating are dealt with. The new process of low-temperature welding of cast iron is explained as well as bronze welding and its many uses.

MINERAL RESOURCES

(Continued from p. 37 A)

Kent Coalfield, Regional Survey Report. (Ministry of Fuel and Power, 1945 : H.M. Stationery Office).

North Eastern Coalfield, Regional Survey Report. (Ministry of Fuel and Power, 1945 : H.M. Stationery Office).

North Western Coalfields, Regional Survey Report. (Ministry of Fuel and Power, 1945 : H.M. Stationery Office).

North Midland Coalfield, Regional Survey Report. (Ministry of Fuel and Power, 1945 : H.M. Stationery Office).

The Coalfields of the Midland Region, Regional Survey Report. (Ministry of Fuel and Power, 1945 : H.M. Stationery Office).

Northumberland and Cumberland Coalfields, Regional Survey Report. (Ministry of Fuel and Power, 1945 : H.M. Stationery Office).

New Canadian Ore Deposit Has High Iron Content. (Steel, 1946, vol. 118, Feb. 25, p. 102). Brief geological details are given of a bed of hematite which has recently been discovered in the Algoma mining area to the north of Lake Huron. The ore contains about 65% of iron.

The Australian Iron and Steel Ltd. Iron Ore Undertaking at Cockatoo Island, Yampi Sound. (B.H.P. Review, 1945, vol. 23, Dec., pp. 1–3). Particulars are given of the discovery, geology, and present state of development of the iron-ore body on Cockatoo Island on the north-western coast of Western Australia. The ore is a high-grade hematite with 69% of iron. Contracts have been placed for diesel-electric power plant, electric shovels, conveyors, and quayside equipment. Living-quarters and buildings for the machinery are already being erected on the island.

Nigeria—Geology and Mineral Resources. F. Dixey. (Bulletin of the Imperial Institute, 1945, vol. 43, Oct.-Dec., pp. 300–311). A short account of the geology and mineral resources of Nigeria is given.

The Formation of Pyritiferous Iron Ores by Addition of Sulphur, and the Nature of Magnetic Pyrites. G. T. Lindroth. (Jernkontorets Annaler, 1946, vol. 130, No. 2, pp. 27–75). (In Swedish). This is the first part of a report on an investigation made in order to study the possibility of removing pyrites, especially magnetic pyrites, from concentrates of pyritiferous iron ores from the leptite system of central Sweden. The geology and crystallographic structure of the ores are dealt with in detail. An outline of the course of the reactions, probably complicated and little known, by which the sulphides were formed is given. They were probably formed by the addition of sulphur from the action of H_2S under high pressure.

French North Africa. A. H. Steedman, D. Gill, and A. Choubersky. (Ministry of Supply, Report of Minerals Division, British Economic Mission, Sept., 1945). This Report is presented by the Minerals Division of the British Economic Mission in Algiers. It contains information and views gained during a period of nearly three years' intensive work in North Africa when large quantities of phosphates and iron were purchased for shipment to the United Kingdom and the condition of the mines, transport facilities, and shipping ports of Algeria, Morocco, and Tunisia were investigated.

FUEL

(Continued from pp. 57 A–59 A)

An Electrical Analogue of the Flow of Heat in a Regenerator System. K. Heindlhofer and B. M. Larsen. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1798 : Preprint Feb., 1945). A description is given of a relatively simple electrical apparatus which, through the close analogy between the flow of heat and of electricity, enables many complex problems of heat flow to be solved quickly and with satisfactory accuracy. Its application to solve problems relating to the design of regenerators is discussed.

The Moisture Sorption of Coals and Its Relation to Their Coking Quality. (Fuel in Science and Practice, 1946, vol. 25, Mar.-Apr., pp. 42–50). The results of experiments on the "sorptive" power of coal are related to its rank and swelling power. The steady diminution of sorptive qualities with advancing coalification may well result from the gradual reduction in number and size of the capillaries inherent in the raw material. The relation of the sorptive power of cokes to that of the parent coal is such that the hypothesis of a capillary structure affords a simple and logical picture.

Dual System for Cleaning Blast Furnace Gas. (Steel, 1946, vol. 118, Jan. 28, pp. 144–146). A brief description is given of the Dovel plant for cleaning blast-furnace gas ; it is a combination of dry and wet methods.

Some Characteristics of Blast Furnace Gas. O. R. Rice. (Iron and Steel Engineer, 1946, vol. 23, Feb., pp. 67–80). Factors affecting and affected by the quantity, calorific value, and chemical analysis of blast-furnace gas are outlined and discussed, special reference being made to the hydrogen content and to the reliable determination of the moisture in the gas. Extensive data are presented, and basic figures used by the author in calculations relating to the production, condition, and combustion of blast-furnace gas are given.

Conservation of Fuel Oil. C. R. Miller. (Iron and Steel Engineer, 1946, vol. 23, Jan., pp. 115–117). A general discussion of the use of fuel oil and methods of conserving it in industry, with data on the consumption in the United States, is presented.

A Heat-Flow Meter for Use in Furnaces. R. H. Baulk and M. W. Thring. (Iron and Steel Institute, 1946, this Journal, Section I.). The first step towards improving the flow of heat to the charge in melting furnaces is to develop an instrument to measure the value of the heat flow at any point in the furnace under the varying conditions of operation. In particular, it was desired to be able to measure the heat transfer by radiation and convection from the flame to the charge and to the furnace walls.

The work on such an instrument has reached a stage where a practical research tool has been evolved and used to compare the heating power of different flames, to examine the effect of flame thickness upon the heat flow from the flame, and to compare the heat flow at different points along and across the flame. It has shown, for example, that measurement of the downward heat flow below the flame at its two ends gives a valuable quantitative measure of the effective flame length; in the case of very slow combustion the downward heat flow at the outgoing end is greater than at the ingoing end. It has also been shown that the heat-absorbing area of the surface of the charge is not very fully utilized either in the steel furnace or in the glass tank, the former because the flame does not cover the whole area and the latter because the cold batch is not spread out.

PRODUCTION OF IRON

(Continued from pp. 59 A–60 A)

Blast Furnace Operation under Elevated Top Pressure. F. Janecek. (Iron and Steel Engineer, 1946, vol. 23, Feb., pp. 88–94, 101). The structural alterations which were made to a blast-furnace with hearth 27 ft. in dia., bosh 30 ft. in dia., and a volume of 46,230 cu. ft. above the tuyeres, to make it suitable for operating at an increased top pressure of 10 lb./sq.in. are described, and the experience gained when working at this pressure is discussed.

Blast-Furnace Theory. I. Reduction and Carburisation of Iron in the Solid State and Carbon Deposition. II. Combustion of Coal in the Hearth. III. Behaviour of Silicon and Manganese—Physical Principles. IV. Thermal Principles of the Blast-Furnace. F. Mobey. (Iron and Steel, 1945, vol. 18, Nov., pp. 481–486; Dec., pp. 629–633; 1946, vol. 19, Jan., pp. 3–6; Feb., pp. 47–51). In this series of articles the author summarizes some of the conclusions reached by a number of investigators of the blast-furnace process.

Mechanizing the Plugging of Blast-Furnace Tapholes. O. Backlund and R. Grundén. (Jernkontorets Annaler, 1946, vol. 130, No. 2, pp. 76–78). (In Swedish). The advantages of installing a clay gun for plugging blast-furnace tapholes are discussed. An increase of about 10% in the production of iron at the Herräng blast-furnaces with a decrease of 0·5% in the coke consumption was achieved.

Drains 640-Ton Salamander from Stack in an Hour. (Steel, 1946, vol. 118, Feb. 11, pp. 114–116). A description is given of the technique adopted to remove furnace bear from a blast-furnace at Indiana Harbor in 1942. A hole 8 ft. high, 6 ft. wide, and 9 ft. towards the back of the furnace was made in the foundation. Two holes 3 in. in dia. and 7 ft. deep were made from the back of the first excavation. These holes acted as guides for the oxygen lance. A 1-in. dia. line was run from a source of oxygen supply 300 ft. away; this was reduced to ½ in. in dia. and 135 ft. of ½-in. tubing was used in lancing the hole to the still molten bear. In 1 hr. 640 tons of iron were drained off and cast into 2-ton pigs in a previously prepared sand pit.

FOUNDRY PRACTICE

(Continued from pp. 60 A–61 A)

Organization of the Foundry Industry in France. V. Delport. (Foundry, 1946, vol. 74, Feb., pp. 113, 226–228). The present organization of the French foundry industry and the functions of the Office Professionnel de la Fonderie, the Centre Technique des Industries de la Fonderie, and the Association Technique de Fonderie are explained.

Foundry Uses Cupola and Direct Metal. P. Dwyer. (Foundry, 1946, vol. 74, Feb., pp. 104–107, 234–236). An illustrated account is given of the moulding and casting technique employed for making an engine-bed casting of grey iron, 26 ft. long × 11 ft. wide × 8 ft. high, weighing 100 tons.

Weather Affects Chill Depth. E. Wisbrun. (Foundry, 1946, vol. 74, Feb., pp. 109, 174–176). The effect of weather conditions on the casting properties of iron is discussed. It was noted that whilst the depth of chill was practically unaltered by a slight fall of snow at above freezing point on the raw materials, a heavy fall of snow at low temperatures considerably increased the depth of chill. The fluidity of the iron remained the same in both cases.

Controlling Gray Iron Quality. J. E. Linabury. (Gray Iron Founders' Society: Foundry, 1946, vol. 74, Feb., pp. 90, 250–254). The factors contributing to the production of good castings in grey iron are examined and recommendations for the control of quality are made.

Some Initial Results on the Influence of Tellurium as a Chill-Inducing Medium in Cast Iron. (Institute of

British Foundrymen, 1946, vol. 78, Mar. 14, pp. 283–287). Four methods of adding tellurium to cast iron were tested and the factors governing the depth of chill were investigated. Direct additions to the ladle were unsatisfactory, but enclosing the tellurium powder in copper foil or copper tubes and putting these in the cold ladle before tapping proved to be effective. With the powder in suspension in a mould-wash excellent control was obtained, as the wash could be applied to faces where no machining of the casting was necessary. Tellurium powder can also be mixed with the facing sand—a procedure which is useful for inaccessible parts of the mould where the wash cannot readily be applied.

Melting and Casting. E. F. Cone. (Materials and Methods, 1946, vol. 23, Jan., pp. 110–114). Developments in steelmaking and in processes of casting iron and steel in the United States during 1945 are reviewed. Successful results have been obtained with the all-basic open-hearth furnace. The trend towards the partial or complete mechanization of iron foundries is noted, and the application of centrifugal casting to peace-time requirements is studied.

The Age of Green and Dry Sand Molding. J. E. Hurst. (Foundry, 1946, vol. 74, Feb., pp. 97, 208–210). The history of sand moulding in Great Britain is outlined. It is known that open sand moulding was practised in Sussex in 1547. Closed sand moulds began to be used in the early part of the eighteenth century.

Mold Cavity Variation. R. Cleland, jun. (Foundry, 1946, vol. 74, Feb., pp. 91, 220–222). Factors affecting the size of the mould cavity and the extent to which it is filled by the metal are discussed. The factors dealt with are the ingredients of the moulding material, namely, moisture, fine sand, cornflour, pitch, sea coal, and clay.

Some Aspects of Green Deformation and Sand Toughness in Sand Control. W. G. Parker. (American Foundryman, 1946, vol. 9, Jan., pp. 50–54). Equipment for measuring the deformation and compressive strength of green sand by American Foundrymen's Association methods is described. When the deformation in thousandths of an inch on 2 in. is multiplied by the green compressive strength and the product is multiplied by 1000, the final product is taken as a measure of the toughness of the sand. With increasing toughness number there is usually a decrease in permeability.

Effect of Table Rise (Strain Rate) on Hot Strength of Bentonite Bonded Sands. H. W. Dietert. (American Foundryman, 1946, vol. 9, Jan., pp. 78–79). In order to obtain reproducible results in hot compressive tests on foundry sands, it is necessary that the strain rate should be specified. Tests showing the effect of different strain rates on the hot strength are discussed, and the rate of 1 in./min. is advocated.

Using Synthetic Resin Core Binders. H. L. Gebhardt. (Foundry, 1946, vol. 74, Feb., pp. 111–112, 198–206). The preparation of cores with synthetic-resin binders is discussed with data on the baking and properties of cores with the ingredients mixed in different proportions.

Pattern and Allied Equipment. W. E. Tharp. (American Foundryman, 1946, vol. 9, Jan., pp. 30–37). Recommendations on the design of patterns and on the interchangeability of patterns and other equipment are made.

Some Pattern Shop Methods and Their Advantages to the Foundry. H. W. Turner. (Institute of British Foundrymen: Foundry Trade Journal, 1946, vol. 78, Mar. 21, pp. 315–320). The preparation of patterns and the fixing of cores for small intricate castings are described.

Sand Mold Drying with Propane-Gas Torches. F. C. Rogers. (American Foundryman, 1946, vol. 9, Jan., pp. 76–77). A description is given of a propane torch which was designed for drying the surface of moulds. The torch uses compressed air at 90–105 lb./sq.in. and propane to 15–20 lb./sq.in.

Moulded Pattern Plates. J. Timbrell. (Foundry Trade Journal, 1946, vol. 78, Mar. 14, pp. 289–290). A method of making a moulded pattern plate with one master pattern and a plaster transfer block for casting a number of identical objects is described.

Casting Tuyeres in the Bethlehem Foundry. P. Dwyer. (Foundry, 1945, vol. 73, Dec., pp. 110–113, 234–236).

Employs Simple Rig for Making Blast Furnace Tuyeres. P. Dwyer. (Steel, 1946, vol. 118, Feb. 4, pp. 138–143). The technique developed by the Bethlehem Steel Company for casting bronze blast-furnace tuyeres is described and illustrated.

How to Cast Ingot Molds. P. Dwyer. (Foundry, 1946, vol. 74, Feb., pp. 158, 160). The technique for making a mould to cast ingot moulds for steel ingots measuring $3\frac{1}{2} \times 3\frac{1}{2}$ in. from 24 to 42 in. long is described.

Duplex Pump Castings. R. H. Brown. (Institute of British Foundrymen: Foundry Trade Journal, 1946, vol. 78, Feb. 28, pp. 223–233). Descriptions are given of the melting and moulding equipment and the moulding and casting technique at a jobbing foundry producing pump castings in iron and bronze.

Moulding Large Meehanite Crankshafts. G. Johnstone. (Foundry Trade Journal, 1946, vol. 78, Mar. 7, pp. 261–264). The following methods of casting large crankshafts in Meehanite cast iron are described and discussed: (1) Vertical pouring and vertical cooling; (2) horizontal pouring and vertical cooling; and (3) horizontal pouring and horizontal cooling.

Precision Casting Practice. (Steel, 1946, vol. 118, Jan. 28, pp. 116–118, 150–156; Feb. 4, pp. 129, 166–170). A detailed description of the "lost wax" process of making small and intricate castings of great accuracy in ferrous and non-ferrous metals is described (*see* Journ. I. and S.I., 1944, No. II., p. 39 A).

Shakeout Handles Large Castings. J. L. Yates. (Foundry, 1946, vol. 74, Feb., pp. 102–103, 166). An illustrated description is given of a large shake-out machine at an American foundry; it has a vibrating table with an area of 396 sq. ft., and it can deal with flasks weighing up to 75 tons.

Sealing Porous Castings. P. S. Fuller. (Engineering Materials and Processing Methods, 1945, vol. 22, Dec., pp. 1746–1750). A method of sealing micropores in castings is described. It consists of im-

pregnating the casting with a thermosetting plastic called " Laminac."

Cleanliness and Safety in the Foundry. J. Thomson. (Safety Engineering: American Foundryman, 1946, vol. 9, Jan., pp. 48–49). A powerful vacuum-cleaning unit for cleaning castings at a Chicago foundry is described and illustrated.

Foundry Ventilation. J. M. Kane. (Foundry, 1946, vol. 74, Feb., pp. 86–88, 228–233; Mar., pp. 104–108, 144–150). The problem involved in the removal of excessive heat, dust, smoke, and fumes from the foundry is analysed and ventilating equipment to meet various conditions is described and illustrated.

PRODUCTION OF STEEL

(Continued from pp. 61 A–63 A)

Restoration of Steel Plant in Southern U.S.S.R. I. Andronov. (Metallurgia, 1946, vol. 33, Feb., p. 188). An account is given of some of the rebuilding and repair work which has been completed at iron and steel works in the U.S.S.R. In the Stalino region 9 blast-furnaces, 18 open-hearth furnaces, 47 batteries of coke-ovens, and 3 Bessemer converters have been restored.

British Steel—Today and Tomorrow. E. N. Simons. (Iron and Steel Engineer, 1946, vol. 23, Feb., pp. 64–66). War-time problems in steel-making which have been overcome in Great Britain and some new processes which have been developed are reviewed. One of the latter is a method of slag control which embodies the charging of lime in furnaces using high percentages of molten phosphoric iron.

Statistical Information. (Jernkontorets Annaler, 1946, vol. 130, No. 2, pp. 81–88). (In Swedish). At a recent meeting of the Järnverksförening (Swedish Ironworks Association) a report was presented on the production, imports, and exports relating to the Swedish iron and steel industry. In this paper thirteen tables from the report are reproduced with notes on the present position of the industry.

Open Hearth Charge Ores. C. Denlinger. (Yearbook of the American Iron and Steel Institute, 1945, pp. 20–28). *See* p. 4 A.

Acid Open-Hearth Steel Making. F. S. Perry. (Sheffield Trades Technical Societies: Man and Metal, 1946, vol. 23, Mar., pp. 28–29; Apr., pp. 40–41). A description is given of an acid open-hearth steel furnace, together with a detailed account of the working of a heat.

A Study of the Durability of Steelworks Ingot Moulds. O. Hengstenberg, K. Knehans, and N. Berndt. (Iron and Steel Institute, 1946, Translation Series, No. 269). This is an English translation of a paper published in Stahl und Eisen, 1941, vol. 61, May 15, pp. 489–498, together with the discussion to which it gave rise. (*See* Journ. I. and S.I., 1941, No. II., p. 120 A).

Arc Furnaces. N. F. Dufty. (Iron and Steel, 1946, vol. 19, Mar., pp. 91–96). The design of electric arc furnaces for steelmaking and their equipment are critically reviewed. Whilst modern arc furnaces are well designed both electrically and mechanically, it is suggested that the men operating them have received too little consideration. A maximum size of 30 tons is adhered to in Great Britain, but there appear to be no logical reasons why the capacity should not be increased as has been done in the United States.

Modern Distribution Systems for Steel Mills. W. C. Bloomquist. (Iron and Steel Engineer, 1946, vol. 23, Feb., pp. 81–84, 101). The planning of a system of electricity distribution in an iron and steel works to make provision for future expansion of the plant is discussed, and the system adopted in a particular case is described.

The Removal of Hydrogen from Steel. J. H. Andrew, H. Lee, A. K. Mallik, and A. G. Quarrell. (Iron and Steel Institute, 1946, this Journal, Section I). A systematic investigation of the removal of hydrogen from a number of steels under isothermal conditions has been carried out, and it is shown that, for all steels, the rate of removal in the intermediate range is high, and that, for certain steels, there is also a peak on the removal curve corresponding to transformation in the pearlitic region. The law governing hydrogen removal in the austenitic range has been determined experimentally, and the empirical relationship supported by theoretical treatment. The effect of transformation upon the removal of hydrogen is illustrated, and the effect of specimen size is also dealt with. It is shown that small partial pressures of hydrogen surrounding the steel at 1100° C. exert a considerable retarding effect upon the removal of hydrogen, although, as might be expected, this effect becomes quite small at 650° C.

In the presence of hydrogen, the transformation of a nickel–chromium–molybdenum steel was retarded in the range 600–450° C., but little effect was observed at lower temperatures. Comparison of the rate of diffusion of hydrogen through similar specimens of different steels showed that alloying elements have little, if any, effect upon the diffusion constant. This was confirmed in the austenitic range by the results obtained for the law of hydrogen removal. The evolution of hydrogen on cooling was followed, and it was found that the maximum rate of evolution corresponded to the end of the thermal transformation.

The mechanical properties of a large number of steels have been determined after treatment in hydrogen and nitrogen, and it is shown that the embrittling effect of hydrogen is apparent in all cases where the hydrogen content is 1·0–3·0 c.c./100 g.

In connection with the work on the removal of hydrogen, much evidence has been obtained upon the formation of hair-line cracks under various conditions. It was found that such cracks could appear in the absence of transformation stresses, and that these stresses are important mainly in that they determine

the particular crack distribution which results. Although no cases were observed of internal crack formation in the absence of hydrogen, certain specimens showed no cracks in spite of a high hydrogen content. The view of the hydrogen-rich constituent put forward in a previous paper has been modified, in that it is not now considered to be necessarily austenitic; it is realized that at the moment of crack formation the steel will be in an extremely brittle condition, owing to its hydrogen content. The exact nature of the hydrogen-rich constituent cannot be given, nor can a detailed explanation of hair-line crack formation be expected, until the much wider problem of low-alloy steels in general is more thoroughly understood.

Steel Plant Is Serviced by Modern Water Works. G. E. Stedman. (Steel, 1946, vol. 118, Feb. 25, pp. 116–121). A description is given of the water-supply service at the new steelworks at Fontana, California. About 2½ million gal./day are supplied by a separate company and two 800-ft. wells have been constructed as a stand-by source of supply. A high degree of recirculation is practised owing to the lack of water in this district.

Lighting Improvements for Steel Plants. D. H. Tuck. (Iron and Steel Engineer, 1946, vol. 23, Feb., pp. 95–100). Improvements in the electric lighting equipment for steelworks which were developed in the United States during the last few years are described and discussed. Methods of measuring lighting efficiency and analysing lighting costs are explained.

FORGING, STAMPING, AND DRAWING

(Continued from p. 64 A)

Some Observations on the Design of Forged Parts. W. Naujoks. (Steel Processing, 1946, vol. 32, Jan., pp. 31–34, 42). Recommendations on the design of forged parts are made with a view to facilitating the forging operations.

Forging Die Design. J. Mueller. (Steel Processing, 1946, vol. 32, Jan., pp. 39–42). The design of dies and rolls for the gap-rolling of shaped and tapering bars is discussed. In gap-rolling, or roll-forging, the die shape takes up about half of the periphery of the rolls, and the billet or bar to be forged is inserted in the gap or recess which is left in the remaining portion of the rolls.

Developments in Production Forging. E. Simister. (Metallurgia, 1946, vol. 33, Feb., pp. 173–177). Descriptions are given of modern drop hammers, counterblow hammers, press-forging machines, and upsetters. The type of production for which each of them is suitable is explained.

Processing and Fabrication of Stainless Steel Sheet and Plate Products. H. S. Schaufus. (Steel Processing, 1946, vol. 32, Jan., pp. 43–46, 59). Data on the effect of different degrees of cold-rolling on the mechanical properties of stainless steel are presented and discussed, and the drawing, spinning, and bending of this material are considered.

Forging and Pressforming. T. C. Du Mond. (Materials and Methods, 1946, vol. 23, Jan., pp. 115–119). Advancements in the forging, press-forming, and stamping of metals in the United States during 1945 are reviewed.

Statistical Quality Control of Metal Stampings. G. R. Armstrong. (Iron Age, 1946, vol. 157, Feb. 21, pp. 46–48). The use of frequency diagrams, correlation charts, and control charts for the statistical control of the quality of metal stampings is explained.

The Witter Process for the Manufacture of Shell Forgings and the Spinning Process for the Manufacture of Bombs. J. L. Johnson. (Yearbook of the American Iron and Steel Institute, 1945, pp. 29–50). A description is given of the Witter cross-roll process of manufacturing closed-end shell forgings to the required length and thickness from a pierced blank; the cavity is so accurately finished that no machining is required. Details of a spinning process for making bombs are also given. In this case the only machining required is that for making the screwthreads on the nose and tail.

Difficulties and Developments in Deep Drawing and Pressing. J. D. Jevons. (Journal of the Institution of Production Engineers, 1946, vol. 25, Mar., pp. 67–103). The causes of "ears," strain-age-embrittlement, stretcher-strain markings, and critical strain crystal growth in deep-drawing and pressing, and methods of dealing with these difficulties are discussed.

Wax Drawing Compound. E. A. Bunting. (Steel, 1946, vol. 118, Feb. 11, pp. 88–89, 138). Details are given of the advantages of using a wax compound as a lubricant for deep-drawing steel, stainless steel, aluminium, and brass.

ROLLING-MILL PRACTICE

(Continued from pp. 64 A–65 A)

Rolling, Wire Drawing, Extruding. H. A. Knight. (Materials and Methods, 1946, vol. 23, Jan., pp. 120–122). The improvements in rolling, wire-drawing, and extruding process which were made known in the United States during 1945 are briefly reviewed. Roller bearings are tending to displace sleeve bearings in rolling-mills. The Krause and Uni-Temper types of mill are described.

Report on die Staatliche Materialprüfungsanstalt an der technische Hochschule at Cannstatterstr. 212, Stuttgart, Germany. (Combined Intelligence Objectives Sub-Committee, File No. XXV-24, 1945: H.M.

Stationery Office). A report is presented on work being done at the above Institute on addition agents to lubricants for the purpose of increasing bearing loads.

Industrial Grease Application. J. R. LeVally. (Iron and Steel Engineer, 1946, vol. 23, Feb., pp. 85–87). Modern grease guns, pumps and connectors for high-pressure lubrication systems are described and illustrated.

Light Steel Building Products. (Combined Intelligence Objectives Sub-Committee, File No. XXVI-22, 1945: H.M. Stationery Office). An account is given of the position reached in Germany in the manufacture of steel windows and doors, with details of the light sections which are rolled.

Metallurgical High Lights in the Recent Manufacture of Rolled Steel Plates and Sheets in Germany. (Combined Intelligence Objectives Sub-Committee, File No. XXVII-29, 1945: H.M. Stationery Office). Important changes which were made during the war in the manufacture of rolled steel plates and sheets in Germany are reviewed.

HEAT-TREATMENT

(Continued from pp. 65 A–66 A)

Calibration of Pit Furnaces for Airplane Propellers. C. A. Liedholm. (Metal Progress, 1946, vol. 49, Feb., pp. 361–368). A detailed description is given of two experimental heat-treatment furnaces which were designed for obtaining data to establish heat-treating specifications for hollow steel propeller blades for aircraft. The furnaces consist of refractory lined pits about 18 ft. deep. There are two interchangeable metal retorts for the high-temperature hardening furnace, and three interchangeable retorts for the low-temperature stress-relieving furnace. All the retorts are about 3 ft. 6 in. in dia., and they vary in length up to 14 ft. 6 in. to suit the size of the blades. The furnaces are heated by banks of resistance heaters in the brickwork. Equipment for the production of carburizing and protective atmospheres is available. Reports on the test runs with temperature, time, and hardness data are presented.

Heat Treating and Heating. T. C. Du Mond. (Materials and Methods, 1946, vol. 23, Jan., pp. 130–134). Achievements in the fields of heat-treating, hardening, and quenching in the United States during 1945 are reviewed.

How Heat-Treatment Affects High-Strength Irons. C. R. Austin. (Machinist, 1946, vol. 89, Feb. 2, pp. 2326–2328; Feb. 16, pp. 2404–2405). The following heat-treatments applicable to Meehanite cast iron are described: (1) Annealing for stress relief; (2) annealing for improved machinability; (3) quenching for hardness; (4) interrupted quenching; (5) martempering; (6) spheroidizing; (7) flame-hardening.

A Theory of the Mechanics of Carburization. C. R. McCloskey and J. H. Loux. (Steel Processing, 1946, vol. 32, Jan., pp. 53–56). The theory that carburization will not take place without carbon dioxide and/or water vapour being present is advanced, and evidence in support of it is put forward. When small controlled amounts of these oxidizing agents are present in the carburizing medium they oxidize the surface of the work; the oxides are immediately reduced by the overwhelming reducing potential and are activated so that they act as a catalyst in dissociating the carbon monoxide into carbon and carbon dioxide. The nascent carbon is absorbed by the activated iron on the surface forming iron carbide.

Insuring Dimensional Accuracy through Heat Treating. (Steel, 1946, vol. 118, Feb. 18, pp. 90–91, 104). The necessity for proper co-operation between the heat-treating and machining departments is emphasized, and a suitable heat-treatment to ensure dimensional accuracy in the case of a particular finished part is outlined.

Molten Baths for the Wire Industry. F. R. Morral. (Wire and Wire Products, 1945, vol. 20, Oct., pp. 736–741). The literature and patents on the application of salt baths in the wire industry for cleaning and heat-treatment are reviewed.

Handling Equipment for Heat-Treatment Furnaces. H. D. Hendrick. (Machinery, 1946, vol. 68, Mar. 14, pp. 329–333; Mar. 21, pp. 367–371). Illustrated descriptions are given of mechanical equipment for charging heat-treatment furnaces, and the design of chain and slat conveyors, pusher furnaces, and quenching equipment is considered.

Importance of a Proper Quench. J. McElgin. (Iron Age, 1946, vol. 157, Feb. 21, pp. 50–51). Experience gained in the heat-treatment of shells of 0·55%-carbon steel is described. It was found advantageous to use a stable quenching oil containing a wetting agent so that a fast rate of cooling through the critical range was obtained. To obtain satisfactory hardness, cooling oil was pumped through jets to quench the inside of the shells.

WELDING AND CUTTING

(Continued from pp. 66 A–68 A)

Metallurgical Aspects of High Quality Pressure Welded Joints for Landing Gear. Part I. Influence of Oxygen. L. Fine, C. H. Maak, and A. R. Ozanich. (Metal Progress, 1946, vol. 49, Feb., pp. 350–355). A description is given of the equipment and process for "uniwelding" steel tubes. In this process the tube ends to be butt-welded are held in alignment in a machine; the joint is surrounded by a ring-shaped burner head with a large number of orifices drilled radially in the inner surface. The head is supplied

with oxygen and acetylene and, when the tube ends have been heated to the right temperature, hydraulic pressure is applied to force the ends together. The welding time for a tube with a $\frac{1}{4}$-in. wall is about 2 min. The mechanism of the diffusion of oxygen in welds of this type is explained.

The Reconditioning of War-Damaged Machine Tools. G. Varley. (Proceedings of the Institution of Mechanical Engineers, 1945, vol. 153, War Emergency, Issue No. 4, pp. 109–111). Methods of repairing machine tools which have been damaged by bombs or fire are described and some examples are illustrated. Three methods were used : (1) Gas welding with cast-iron filler rods containing at least 3% of silicon ; (2) electric welding with covered electrodes suitable for cast iron ; and (3) bronze welding by arc using covered electrodes.

Design Considerations for Welded Machinery Parts. G. L. Snyder. (American Society of Mechanical Engineers : Metallurgia, 1946, vol. 33, Feb., pp. 183–188). **Designing Welded Machinery Parts.** G. L. Snyder. (Steel, 1946, vol. 118, Feb. 4, pp. 124–126, 162–164 ; Feb. 11, pp. 98–102, 155 ; Feb. 18, pp. 100–102, 140–144). Factors affecting the design of dynamically-loaded welded machinery parts are discussed.

High-Speed Spotwelding. B. Gross. (Iron Age, 1946, vol. 157, Feb. 21, pp. 52–55). Factors influencing high-speed spot-welding are discussed. A small apparatus for obtaining records of the time-pressure and the time-current relationships is described.

Three-Phase Distribution and Electric Welding and Furnace Loads. A. Langley Morris. (Institution of Electrical Engineers, Mar. 13, 1946). Methods of connecting single-phase welding equipment to 3-phase mains are dealt with. The single-phase transformer and the open-delta series-connected transformer are compared. Mention is made of the power-factor correction of single-phase loads, from which there has been developed a method of obtaining balanced 3-phase/single-phase conversion by static means. Full details of this phase-converting system for arc- and resistance-welding machines are given.

Notes on the Conditions of Fracture of Medium Steel Ship Plates. D. F. Windenburg and W. P. Roop. (Welding Journal, 1945, vol. 24, Nov., pp. 580-S–587-S). Tensile tests were made on notched specimens of steel ship plates measuring $24 \times 12 \times \frac{3}{4}$ in. The notch took the form of a hole through the centre ; in some cases hacksaw cuts $\frac{1}{16}$ in. long were made in opposite sides of the hole transverse to the direction of testing. The plates were tested to rupture at various controlled temperatures and the nature of the fracture was examined. Specimens notched with a hole and hacksaw cuts broke with a brittle fracture at 75° C. whereas a similar specimen broke with a ductile fracture at 100° F. The characteristics of the fractures were not altered by sharp fatigue cracks at the ends of the saw cuts.

Report on Welding in German Shipyards. (Combined Intelligence Objectives Sub-Committee, File No. XXXII-100, 1945 : H.M. Stationery Office).

Fatigue Properties of Flash Welds. H. J. Grover, R. W. Bennett, and G. M. Foley. (Welding Journal, 1945, vol. 24, Nov., pp. 599-S–617-S). An investigation of the fatigue properties of flash-welded chromium-molybdenum steel, conducted at the Battelle Memorial Institute is described in three parts dealing respectively with welded bars, plates, and tubes. In Part I. the properties of various zones of flash-welded bars were studied by rotating-beam tests on small specimens. Part II. describes bending-fatigue tests on flash-welded plates to determine the effect of weld reinforcement. In Part III. the effect of surplus metal on the inside of the tube (" internal flash ") was tested in the rotating-beam machine.

Some Causes of Brittle Failures in Welded Mild Steel Structures. H. E. Kennedy. (Welding Journal, 1945, vol. 24, Nov., pp. 388-S–398-S). Factors affecting the formation of brittle fractures in welded steel structures were investigated by notched-bar tensile tests, bend tests, and concentrated load tests. Changes in metallurgical structure caused by temperature changes were found to be the most important factor contributing to brittle fracture. Of the measures calculated to prevent brittle fracture, post-heating to 1100° F. was found to be the best, and preheating to 400° F. or higher was very beneficial.

The Straining of Deposited Weld Metal During Cooling. E. P. De Garmo. (Welding Journal, 1945, vol. 24, Nov., pp. 561-S–563-S). An investigation of the plastic flow that takes place in a weld as the result of cooling is described. In a butt weld by the Union-melt process, exceeding 20 in. in length, the weld metal half-way along the weld is subjected during cooling to a plastic strain greater than that corresponding to the yield point of the material.

Weldability Tests of Cast Steels. F. S. McKenna and C. E. Jackson. (Welding Journal, 1945, vol. 24, Nov., pp. 573-S–579-S). The T-bend and nick-bend tests are described and recommended as direct quantitative tests for determining the effect of changes in welding technique and heat-treatment on the ductility of cast steels, and for indicating the relative weldability of cast steels of any composition.

Oxyacetylene Cutting Tactics for Various Alloy Steels. G. V. Slottman. (Machinery, 1946, vol. 89, Mar. 16, pp. 2542–2544). Techniques for the oxy-acetylene cutting of alloy steels are described. High-speed steel, high-nickel steel, stainless steels with 12–20% of chromium and 7–10% of nickel, and cast iron are difficult to cut. A reciprocating motion of the torch is required for all of these except cast iron which is cut with an oscillating motion to make a wide cut.

CLEANING AND PICKLING OF METALS

(Continued from pp. 47 A–48 A)

Sodium Hydride Descaling. W. B. Billingsley. (Canadian Metals and Metallurgical Industries, 1946, vol. 9, Jan., pp. 19–24). The sodium-hydride pickling process has been described by L. W. Townsend (*see* p. 48 A). In the present paper the full-scale plant at the works of the Rustless Iron and Steel Corporation

is described and illustrated, and the advantages of the process are pointed out.

Three Examples of Modern Practice in the Recovery of Waste Pickling Liquors. P. De Lattre. (Sheet Metal Industries, 1946, vol. 23, Feb., pp. 282–286, 288). Descriptions are given of three pickling plants belonging to a tube manufacturing company in Belgium. At two of these plants the spent liquor is treated as a whole, which involves the use of rather large deposit and settling tanks. In the third, where high-precision tubes are made, a continuous regeneration plant has been installed eliminating entirely the need for settling and deposit tanks.

The Chemistry and Mechanism of Steel Pickling. B. N. Reavell. (Institute of Vitreous Enamellers: Sheet Metal Industries, 1946, vol. 23, Feb., pp. 323–326). The mechanism of pickling is explained on the basis of electrochemical rather than chemical reactions. The following nett equations may be taken to summarize the reactions involved:

$$Fe_3O_4 + Fe + 4H_2SO_4 = 4FeSO_4 + 4H_2O$$
$$Fe_3O_4 + Fe + 8HCl = 4FeCl_2 + 4H_2O.$$

As long as scale remains in contact with the base metal, then, in the presence of the electrolyte, innumerable electrolytic cells of the types $Fe–Fe_2O_3$, $Fe–FeO–Fe_3O_4$, $Fe–Fe_3O_4–Fe_2O_3$ are set up; these lead to a reduction of some of the oxides to a ferrous state which reacts with the acid ions to form ferrous salts which pass into solution.

Batch and continuous plants are briefly described with notes on acid neutralization and acid recovery.

The Effect of Various Surface Treatments in Cleaning and Preparing Copper, Nickel, and Steel for Chromium Plating. W. T. Tucker and R. L. Flint. (Electrochemical Society, Oct., 1945, Preprint No. 28). The Hull cell was utilized as a device for studying the reluctance of certain cathodes to accept a chromium deposit. Results of excellent reproducibility were obtained. It was concluded that the commercial methods of surface preparation evolved since the inception of chromium plating were adequate if certain limiting factors were duly considered. Neither copper nor steel were found to be acutely sensitive to methods of surface preparation.

Pickling of Sheet Steel prior to Vitreous Porcelain Enamelling. F. Stanley. (Australasian Engineer, 1946, Jan. 7, pp. 64–69). Detailed recommendations on pickling procedure for steel sheet which is to be vitreous-enamelled are made and formulæ for the rapid calculation of acid additions are presented. "Pickle pill" and titration methods of bath control are described.

Cleaning and De-Whiskering Nails. A. J. Lehman. (Steel, 1946, vol. 118, Feb. 18, pp. 123–126). In the nail-manufacturing process the nails come from the dies coated with a mixture of lime and soap, and there is also a small "rag" of metal left at the point by the knives. A new process for removing both coating and rag is described. The nails are tumbled in a hot solution and sawdust is used only to dry and burnish the clean nails. The continuous unit has automatic charging and discharging devices.

COATING OF METALS

(Continued from pp. 48 A–50 A)

Metallic Coating by Immersion. E. Dupuy. (Revue de Métallurgie, Mémoires, 1944, vol. 41, July, pp. 208–216). The author discusses the use of various metals, such as tin, zinc, copper, lead, cadmium, and aluminium, for coatings for one of the following purposes: (1) To modify the external appearance; (2) to resist corrosion; (3) to facilitate cold-working; and (4) for welding and soldering. A brief description is given of immersion methods of coating with tin, zinc, lead, cadmium, and aluminium.

Precision Seamless Tubing by Electrodeposition. J. Albin. (Iron Age, 1946, vol. 157, Jan. 10, pp. 54–58). Some details of the Ekko process of plating iron have been given in an earlier publication (*see* Journ. I. and S.I., 1940, No. I., p. 155 A). In this article particulars are given of the application of the process to the manufacture, with a high degree of accuracy, of seamless tubes of irregular shape.

The Electrodeposition of Nickel-Cobalt-Tungsten Alloys from an Acid Plating Bath. P. F. Hoglund and M. L. Holt. (Electrochemical Society, Oct., 1945, Preprint No. 24). The electrodeposition of nickel-cobalt-tungsten alloys from an aqueous plating bath is reported. The most satisfactory bath was found to be a modified Fink and Lah borate-type of nickel-cobalt bath to which was added a small amount of sodium tungstate.

Wire Galvanizing—Herman Process. R. Hussey. (Wire and Wire Products, 1945, vol. 20, Dec., pp. 927–931). A description is given of the Herman hot-dip process for galvanizing wire. In this process there is a small auxiliary zinc pot at the exit end of the main pot. The wires are drawn up through a slot-shaped nozzle, 10 in. high, through which molten zinc is circulated. Above the nozzle there is a chamber of about 6 cu. ft. which is kept full of natural gas. The wires pass up through the gas and are cooled in a stream of water directed across them.

The Formation of Dross in Hot-Dip Galvanizing. P. M. Fisk and F. F. Pollak. (Sheet Metal Industries, 1946, vol. 23, Mar., pp. 490–492, 494). The effect of changes in the pickling and washing conditions on the amount of free iron in mg./sq.m. produced on specimens of steel plate $4\frac{1}{2} \times 1\frac{3}{4}$ in. was investigated. As 3 lb. of iron produce 100 lb. of dross in the subsequent hot-dip galvanizing process, the data obtained are important in studying methods of reducing dross formation. The amount of free iron increased rapidly with increasing roughness of surface and with the quantity of iron in the pickle solution. Steel pickled in cold hydrochloric acid is not freed from iron salts under ordinary production conditions after dipping in static water as easily as the steel pickled in sulphuric acid.

The Electrodeposition of Tin-Zinc Alloys. R. M. Angles. (Journal of the Electrodepositors' Technical Society, 1946, vol. 21, pp. 45–64). The results of an investigation on the deposition of tin-zinc alloys, ranging in composition from 92/8 to 28/72, are discussed, and the procedure for depositing the alloy containing 78% of tin, which appears to be the most serviceable, is given.

The Protective Value of Lead and Lead-Tin Deposits on Steel. A. H. Du Rose. (Electrochemical Society, Oct., 1945, Preprint No. 7). The effects of the thickness of the deposit and of the concentration and nature of the addition agents to the plating bath on the protective value of lead and lead-tin coatings on steel were investigated. An increase in the glue concentration in the lead fluoborate plating solution gives better covering and throwing power. Lead deposits give a surprisingly good protection to steel against corrosion. A lead-tin alloy deposit containing 5–6% of tin gives better protection than either pure lead or an alloy higher in tin.

Metallizing in Welding Fabrication. W. J. Brooking. (Canadian Metals and Metallurgical Industries, 1946, vol. 9, Jan., pp. 14–18, 24). The application of metal-spraying for the repair of worn shafts, hubs, and bushes, as well as for the salvage of over-machined parts, is described and illustrated.

The Application of Cold Metal Spraying. V. G. Young. (Diesel Engine Users' Association: Australasian Engineer, 1946, Jan. 7, pp. 61–62). The advantages and limitations of the metal-spraying process for the repair of diesel engine parts are discussed.

Electrical Testing of Wire, etc. (Wire Industry, 1946, vol. 13, Feb., pp. 103–105). Descriptions are given of an enamelled wire pin-hole tester and a magnetic sorting bridge. The former instrument counts the number of flaws and pin-holes of a given maximum resistance in a length of enamelled wire. The sorting bridge enables samples of any ferromagnetic materials to be tested and determinations made of differences in their composition, hardness, temper, and electrical conductivity.

Conveyorized Enamelling of Castings. H. Chase. (Iron Age, 1946, vol. 157, Feb. 14, pp. 53–55). A description is given of a highly mechanized plant at which cast and stamped components are cleaned, bonderized, sprayed, and baked on a continuous chain conveyor with specially designed controls.

The Decoration of Blackplate and Tinplate. T. A. Miller. (Sheet Metal Industries, 1946, vol. 23, Feb., pp. 291–292, 298). Recent developments in the sizing, lacquering, and printing for the decoration of blackplate and tinplate are reviewed, and some future possibilities discussed.

Synthetic Coatings for Gasoline Tanks. (Combined Intelligence Objectives Sub-Committee, File No. XXX-67, 1945: H.M. Stationery Office). A brief report on German methods of applying synthetic coatings on the inside of tanks used for storing petrol is presented. For small tanks, baked phenol-formaldehyde resin coatings were used, whereas for larger tanks a plasticized nitrocellulose air-dried lacquer was applied. This lacquer, however, was modified by the addition of soluble urea-formaldehyde resin which was cured by the addition of phosphoric acid to effect good adhesion and to render the film insoluble in petrol.

Plastic Skin Protective Coatings Convert to Peace-Time Applications. B. Gould. (Corrosion and Material Protection, 1946, vol. 3, Feb., pp. 6–9, 12). Some particulars are given of a plastic coating with a cellulose acetate butyrate base into which steel parts can be dipped. A bath kept at 300–340° F. is used and, on withdrawing the part, the solution solidifies leaving a transparent film completely impervious to moisture.

PROPERTIES AND TESTS

(Continued from pp. 68 A–69 A)

Young's Modulus—Its Metallurgical Aspects. D. J. Mack. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1936: Metals Technology, 1945, vol. 12, Dec.). The literature on Young's modulus is critically reviewed, and a number of empirical relations correlating the modulus with other properties such as atomic volume and specific heat are considered; the best is considered to be that of J. P. Andrews. The bibliography contains 103 references.

Correlation of Inspection Methods in the Metal Industry. J. V. Rigbey. (Canadian Metals and Metallurgical Industries, 1945, vol. 8, Nov., pp. 36–40; Dec., pp. 38–41, 48–52). Modern methods of testing and inspecting metals are reviewed and their significance is discussed.

Sulphur—Effects in Cast Iron and Steel. E. S. Brown. (Sheffield Trades Technical Societies: Iron and Steel, 1946, vol. 19, Feb., pp. 71–74). The effects of sulphur on the properties of cast iron and steel are discussed, and data on the reduction in tensile properties of steel at high temperatures with increasing sulphur content are presented. Some experiments on desulphurization with soda ash are described; the conclusions derived from them are: (1) The amount of soda ash used should be related to the initial sulphur contents; (2) it is uneconomical to make additions of more than 22½ lb. of soda ash per ton of steel; and (3) a double treatment is essential when an initial sulphur content exceeding 0·090% is to be brought down to 0·010–0·020%.

Shot-Peening. (Aircraft Production, 1945, vol. 7, Oct., pp. 478–480). The characteristics of the shot used for shot-peening and data on the increase in fatigue strength obtained by peening wire are discussed. A method of checking the depth to which the peening affects the steel is described.

The Hardenability Concept. J. H. Hollomon and L. D. Jaffe. (American Institute of Mining and Metallurgical Engineers, Technical Publication No.

1926: Metals Technology, 1946, vol. 13, Jan.). Experimental determinations of the hardenabilities of some high-alloy steels compared with the hardenability calculated by the Grossmann system have indicated that the calculated hardenabilities are too high. The Grossmann system and the whole concept of hardenability are therefore examined in this paper. In order to determine the effects of the alloying elements on the bainitic hardenability, the alloying elements should be added to the steels in which bainite limits the formation of martensite. This occurs in the case of steels containing about 0·25% of molybdenum. Since no carefully planned experiments of this type have been performed, a tentative system is suggested as an interim measure to permit very rough estimates of the bainitic hardenability to be made. Hardenabilities calculated by this system for a number of steels have been found to check well with the measured hardenabilities. One of the conclusions based on the results of isothermal experiments in this tentative system is that molybdenum and the grain size of the austenite have little, if any, effect on the rate of bainite formation.

Hardenability and Quench Cracking. L. D. Jaffe and J. H. Hollomon. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1927: Metals Technology, 1946, vol. 13, Jan.). A method is suggested for selecting the composition of alloy steels which will have a specified hardenability with a minimum quench-cracking tendency.

Progress Report on Hardenability Bands. (Yearbook of the American Iron and Steel Institute, 1945, pp. 51–84). This Report was prepared by the Technical Committee on Alloy Steel of the American Iron and Steel Institute. It deals with the advantages and disadvantages of specifying alloy steels by hardenability instead of by analysis. The production records for the principal alloy steels manufactured in a period of two months in the United States show that 14% of the open-hearth heats and 4·4% of the electric-furnace heats were outside of the specified limits of analysis. Frequency curves for the contents of various elements in these "off" heats are presented and discussed. The general conclusion is that it would be both practicable and advantageous to widen the present limits of specified contents of elements and to specify hardenability limits based on Jominy end-quench tests.

Hardenability. A. Portevin. (Revue de Métallurgie, Mémoires, 1944, vol. 41, July, pp. 193–203; Aug., pp. 243–248). The author discusses the theoretical and practical aspects of hardenability of steel and cast iron. Methods of testing hardenability are described and the value and application of the results obtained are explained.

New Developments in Hardness Testing. V. E. Lysaght. (Metallurgia, 1946, vol. 33, Feb., pp. 210–212). The Knoop hardness-testing indenter, the Tukon tester for applying light loads with the Knoop indenter, and some work carried out with them are described.

Applications of Eberbach Microhardness Tester. H. E. Bernhardt. (Steel, 1946, vol. 118, Feb. 25, pp. 92–100, 140). The Eberbach microhardness testing instrument is described. It operates with a spring-loaded diamond indenter of similar shape to that in the Vickers machine. Results obtained with it are presented and compared with those obtained by other methods.

The Value of the Impact Test in the Study of Temper Brittleness. H. Jolivet and G. Vidal. (Revue de Métallurgie, Mémoires, 1944, vol. 41, Nov., pp. 378–388; Dec., pp. 403–408). The results of impact tests on specimens from eleven heats of 1·3–2·0%-chromium 0·26%-molybdenum 0·23%-vanadium steel are presented. Mesnager notched specimens subjected to various heat-treatments were used. The influence of the deformation velocity, temperature, notch shape, and heat-treatment were investigated. Temper brittleness can be detected only by measuring the impact strength when the test is carried out under favourable conditions.

The Physical Nature and Properties of Modern Magnet Steels and the Practical Design and Testing of Permanent Magnets. E. Spahn. (Schweizer Archiv, 1944, vol. 10, Oct., pp. 313–322).

Testing and Inspection. H. A. Knight. (Materials and Methods, 1946, vol. 23, Jan., pp. 145–150). The methods of testing and inspection which are reviewed in this paper include electronic measurements, spectrography, magnetic inspection, hardness testing with light loads, strain gauge determinations, the measurement of surface roughness, and radiography.

Other Non-Destructive Methods of Testing. J. W. Juppenlatz. (American Foundryman, 1946, vol. 9, Jan., pp. 38–41). Non-destructive methods of testing castings other than radiography and magnetic particle testing are described and discussed. These include: (*a*) Washing the surface with a fluorescent penetrating liquid and subsequent examination under ultra-violet light; (*b*) static loading tests in which the deflection is measured; (*c*) air-pressure and liquid-pressure tests; and (*d*) supersonic tests.

Crack Inspection. F. W. Giles. (Engineering Inspection, 1945, vol. 10, Winter Issue, pp. 22–26). Methods of detecting cracks in metals are briefly reviewed. Each method has a particular merit against the others in a given circumstance and experience proves that there is room for them all.

Electronic Testing. (Iron and Steel, 1946, vol. 19, Mar., pp. 119–120). Electronic instruments for comparing the mechanical characteristics of great numbers of parts, for detecting cracks, and for determining the thickness of non-magnetic coatings on steel are described.

Metallic Creep. (Engineer, 1946, vol. 181, Feb. 15, pp. 148–149; Feb. 22, pp. 170–171). **Creep of Metals.** (Engineering, 1946, vol. 161, Mar. 8, pp. 233–235; Mar. 15, pp. 258–259). A brief report is presented of the proceedings at a conference on metallic creep which was held in London on February 5, 1946. It was organized by the Department of Scientific and Industrial Research and about 150 representatives of research establishments, universities, and industrial organizations were present.

A Comparison of Some Carbon Steels on the Basis of Various Creep Limits. A. E. Johnson and H. J. Tapsell. (Proceedings of the Institution of Mechanical

Engineers, 1945, vol. 153, War Emergency Issue No. 6, pp. 169–179). This paper describes work done to determine whether any relationship exists between the results of the various short-time and long-time creep tests, which will permit the use of short-time tests, not merely as a means of separating good from bad steels, but also as the basis of design stresses intended to give satisfactory performance over working periods of considerable duration. Indications are given of some of the difficulties and limitations in the use of short-time creep results in the case of carbon steels, and suggestions are made for some rational basis for steps which might be taken to offset these difficulties and limitations by applying the results of short-time tests to the estimation of design stresses.

Creep Properties of Steels Utilized in High-Pressure and High-Temperature Superheater and Steam Pipe Practice. Part II. 0·5 Per Cent Molybdenum Steels. H. J. Tapsell and R. W. Ridley. (Report J/T 136 of the British Electrical and Allied Industries Research Association: Proceedings of the Institution of Mechanical Engineers, 1945, vol. 153, War Emergency Issue No. 6, pp. 181–192). This report deals with the creep properties of carbon-molybdenum steels in the form of a superheater header, superheater tube, and steam pipe manufactured for service at temperatures above 450° C. The investigation was carried out in a similar manner to that described in Part I. (*see* Journ. I. and S.I., 1944, No. II., p. 88 A), which dealt with carbon steels and, as in the former case, the object was to obtain data for the estimation of the stress-temperature relationships for from 0·1 to 0·5% creep in various periods up to 100,000 hr. It is considered that the data provide a satisfactory basis for design.

Valve Facing Alloy Resists Corrosion at High Temperatures. V. C. Young. (Engineering Materials and Processing Methods, 1945, vol. 22, Dec., pp. 1744–1745). Some particulars are given of the properties of an alloy called "Eatonite," containing chromium, nickel, tungsten, and cobalt, which maintains a high hardness at red heat, and is used for facing the valves of aero-engines.

16-25-6 Alloy for Gas Turbines. M. Fleischmann. (Iron Age, 1946, vol. 157, Jan. 17, pp. 44–53; Jan. 24, pp. 50–60). A comprehensive account is given of the development by the Timken Roller Bearing Company of a 16/25/6 chromium-nickel-molybdenum steel for use at high temperature in gas turbines. Details are given of its microstructure, heat-treatment, and room-temperature and high-temperature properties.

The Influence of Heat-Treatment and Micro-Structure of Austenitic Heat-Resisting Steels on the Scatter of Creep-Test Results. E. Morlet. (Revue de Métallurgie, Mémoires, 1944, vol. 41, June, pp. 161–168; Sept., pp. 284–291; Oct., pp. 346–352). The author presents a report on a study of creep resistance; it is in three parts which deal with: (1) Creep-testing machines and the accuracy of the measurements; (2) the influence of heat-treatment; and (3) the influence of the grain size on creep tests of short and long duration. Creep tests were made at 650°, 750°, and 850° C. on an 18/8 austenitic chromium-nickel steel with an addition of 3·5–4·0% of tungsten, and on an austenitic-ferritic 24/12 chromium-nickel steel. The effect of heat-treatment and of different grain sizes on the creep resistance, creep rate, and scatter of the results was examined. The following conclusions were arrived at: (1) In order to limit the scatter of the creep-test results, the heat-treatment should be carried out at the following temperatures: Quenching from 1150° C. followed by tempering at 900° C. for 6 hr. (2) A coarse-grained steel possesses a higher creep resistance, and it is possible to represent by a diagram on a logarithmic scale the variation of the creep rate in relation to the number of grains per square millimetre. (3) The scatter of the results is more pronounced in creep tests of long duration, and this shows the superiority of coarse-grained forged steel over a fine-grained rolled steel. (4) The coarse grain, caused by overheating, makes the steel susceptible to intergranular fractures which considerably affect the creep resistance.

Report on Metals for Elevated Temperatures. (Combined Intelligence Objectives Sub-Committee: File No. XXXII-99, 1945: H.M. Stationery Office). Information is given on German methods of creep-testing, their alloy steels for boilers, and on the Krupp heat-resisting steels.

Report on Materials for Shipbuilding Applications. (Combined Intelligence Objectives Sub-Committee, File No. XXXII-105, 1945: H.M. Stationery Office).

The Diffusion of Hydrogen Through Iron at Room Temperature. J. H. De Boer and J. D. Fast. (Recueil des Travaux Chimiques des Pays-Bas, 1939, vol. 58, Sept.-Oct., pp. 984–993). When iron sheets are brought into contact with aqueous solutions of definite *p*H values, diffusion of hydrogen through the iron takes place in all cases where the liquid dissolves iron with the evolution of hydrogen. No diffusion takes place if the *p*H value of the liquid is such that there is no evolution of nascent hydrogen. This proves that atoms of hydrogen formed by the chemical reaction may enter the iron, whilst the hydrogen ions contained in the solutions are unable to do so. In accordance with these results it was found that hydrogen atoms formed in hydrogen gas by means of an incandescent tungsten coil readily enter iron plates at room temperature. An iron vessel filled with hydrogen can be evacuated in this way. As hydrogen can diffuse only through the iron from the side where atoms are produced, it accumulates on the other side of the iron wall and diffusion takes place against an atmosphere of hydrogen at any pressure.

A Test for Hydrogen Embrittlement and Its Application to 17 per Cent Chromium, 1 per Cent Carbon Stainless-Steel Wire. C. A. Zapffe and M. Eleanor Haslem. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1954: Metals Technology, 1946, vol. 13, Jan.). An apparatus is described with which hydrogen embrittlement in wire can be measured as a function of the angle at which breaking occurs during a single bend made at constant speed. This simple instrument is used first to get an approximate indication of the sensitivity of various grades of stainless steel wire to hydrogen embrittlement and then to study factors controlling the absorption of hydrogen during cathodic and acid pickling and the evolution of that gas during subse-

quent recovery. After establishing the fundamental relationships of charging time, temperature, and current density, it was found that many inhibitors, often advertised as inhibiting or preventing hydrogen embrittlement, actually caused a marked increase in embrittlement when compared with the acid bath without such additions. Only two reagents, neither of which was a commercial product, were found to be true inhibitors at certain concentrations; at other concentrations they, too, aggravated embrittlement. The peculiarities of a group of these reagents were studied in detail. During the recovery period the specimens first decreased in ductility, a very marked spontaneous embrittlement being observed. As an example of the effect of temperature on the rate of recovery, it was noted that, after 40 sec. at 180° C., the recovery was the same as after 71 hr. at 0° C. The evidence obtained in these tests supported the " planar-pressure " theory of hydrogen embrittlement.

Steels and Irons. F. P. Peters. (Materials and Methods, 1946, vol. 23, Jan., pp. 86–89). Some of the ferrous materials developed during 1945 are reviewed with notes on their properties. They include alloys for use at high temperature, high-ductility steels for forming, boron-bearing malleable cast iron, " H " steels with specified limits of hardenability, steels with special magnetic properties, and tool steels.

Railway Rails. M. Cambournac. (Revue de Métallurgie, Mémoires, 1944, vol. 41, Aug., pp. 233–242; Sept., pp. 301–312). The author presents the results of the investigations made by a special Commission for Rails in France, whose task was to examine the causes of various defects and to suggest means of eliminating them.

Some Interesting Properties of Titanium Steels. G. F. Comstock. (Journal of the American Ceramic Society, 1946, vol. 29, Jan., pp. 1–7). Steel containing sufficient titanium to combine with all of the carbon does not react with vitreous enamels like ordinary low-carbon steel or ingot iron; white cover-coat enamels, without a ground coat, can therefore be fused on it without blistering or the formation of black specks. The titanium content must be more than $4\frac{1}{2}$ times the carbon content. This steel has been made commercially by the basic open-hearth process, and the requirements for successful manufacture are explained. It is of excellent quality for deep-drawing and does not have a yield point, even when normalized or annealed, so that stretcher-strains cannot occur in it. It is not subject to strain-ageing even when strained as much as 17% and aged at 450° F.

Steel Wool—Characteristics of Materials and Manufacture. L. E. Browne. (Steel, 1946, vol. 118, Feb. 25, pp. 87, 124). The manufacture of steel wool is described and its properties discussed. Hot-rolled steel (containing about 0·12% of carbon and 0·90% of manganese) $\frac{7}{32}$ in. in dia. is drawn down to 0·01 in. wire, and the fine strands for making into wool are shaved off the wire with a serrated-edged tool which removes several strands at a time. The strands are made in seven sizes between 0·0018 in. and 0·0080 in. in dia. The binding of the strands into wool is not described.

Ferrous Metallurgical Research. N. A. Matthews. (Journal of Applied Physics, 1945, vol. 16, Dec., pp. 780–787). The organization, equipment, and work being done at the Watertown Arsenal Laboratory, Massachusetts, are described and discussed.

A Workable Classification of Toolsteels. C. L. Hibert. (Metal Progress, 1946, vol. 49, Feb., pp. 333–341). The method of coding tool steels adopted by the Consolidated Vultee Aircraft Corporation is described. The steels were divided into five major groups with 14 sub-divisions. The groups, each of which with its own distinctive mark, are: (1) General-purpose tool steels; (2) oil-hardening die steels; (3) silicon-manganese shock-resisting chisel and punch steels; (4) high-carbon high-chromium punch and die steels; and (5) high-speed tool steels.

Nomenclature and Standards. (Jernkontorets Annaler, 1946, vol. 130, No. 2, pp. 79–80). (In Swedish). The *Tekniska Nomenklaturcentral* (Bureau for Technical Nomenclature) is about to publish a list of 200 definitions and terms relating to the strength of materials, which have been agreed upon by a committee of fifteen Swedish scientists and industrialists. A short review of some of the definitions is given and comments are invited.

METALLOGRAPHY AND CONSTITUTION

(Continued from pp. 69 A–70 A)

A Temper-Etching Process for Distinguishing between Ferrite, Austenite, and Carbides in the Structure of High-Chromium Steels. H. Kessner. (Iron and Steel Institute, 1946, Translation Series, No. 270). This is an English translation of a paper which was published in Archiv für das Eisenhüttenwesen, 1942, vol. 16, Oct., pp. 145–146. (*See* p. 15 A).

Industry's Debt to Röntgen. G. L. Clark. (Industrial Radiography, 1945, vol. 4, Winter Issue, pp. 30–34, 49). The development of the use of X-rays for industrial purposes during the last fifty years is outlined.

Some Questions that Can be Answered by X-Ray Diffraction. C. S. Barrett. (Industrial Radiography, 1945, vol. 4, Winter Issue, pp. 12–18). A general survey of the field covered by X-ray diffraction is presented. Its application to obtain data for constructing phase diagrams and studying the effects of deforming and heat-treating metals is dealt with.

Atomic Structure of Steels. Audrey M. B. Parker. (Steel, 1946, vol. 118, Feb. 25, pp. 89, 127). The Debye and Scherrer method of obtaining X-ray powder photographs is described and their significance in studying the atomic structure of steels is explained.

Technique and the Future of Industrial Radiography. M. B. Evans. (Industrial Radiography, 1945, vol. 4, Winter Issue, pp. 19–22). War-time improvements in X-ray technique are reviewed and

an assessment is made of their value in application to peace-time requirements.

Ultra High Speed Radiography. E. R. Thilo. (Industrial Radiography, 1945, vol. 4, Winter Issue, pp. 40–43). The technique for obtaining high-speed X-ray photographs is described with examples of its use for studying the penetration of plates by bullets and other ballistical problems.

Two Million Volt X-Ray Inspection Equipment. D. Goodman. (Metal Progress, 1946, vol. 49, Feb., pp. 327–332).

High-Voltage Radiography. D. Goodman. (Iron Age, 1946, vol. 157, Feb. 21, pp. 58–61). Experience gained from the use of 1,000,000-V. and 2,000,000-V. X-ray equipment in the United States is reported. The latter size of equipment weighs about 5000 lb. and measures 5 ft. in dia. and 11 ft. in length; with it, a radiograph can be taken through steel 12 in. thick at a distance of 3 ft. with an exposure of 2 hr.

Application of X-Ray Analysis to Metallurgy. M. Païc. (Revue de Métallurgie, Mémoires, 1944, vol. 41, June, pp. 169–179). The author discusses the method of quantitative micro-analysis which is based on the absorption of X-rays passing through a substance. He investigates the loss of intensity of X-rays when penetrating various metals, and shows on a diagram the relationship between the absorption and atomic number of various elements. The practical application of this law is shown by several examples.

Report on Radiography and Magnetic Inspection in German Shipyards. (Combined Intelligence Objectives Sub-Committee, File No. XXXII-102, 1945: H.M. Stationery Office).

Microstructure Related to Properties of Cast Iron. W. E. Mahin and H. W. Lownie, jun. (American Foundryman, 1946, vol. 9, Jan., pp. 20–28). Explanations of the terms cementite, ferrite, pearlite, and steadite are given and the manner in which these constituents and graphite affect the properties of cast iron is explained.

Introduction to a Study of Coalescence in Eutectoid and Hypereutectoid Steels. N. T. Belaiew. (Revue de Métallurgie, Mémoires, 1944, vol. 41, Mar., pp. 65–70; Apr., pp. 122–130; May, pp. 150–158; June, pp. 188–191; July, pp. 227–232; Aug., pp. 249–257; Sept., pp. 292–300; Oct., pp. 341–345). A comprehensive study is presented on the influence of carbides on the quality of steel. The importance of three main factors is discussed; those factors are the size, form, and distribution of the carbides. The author's general conclusions are: (1) The quantity, distribution, form, and size of the carbides exert a profound influence on the properties of steel. In the first place, the elastic limit, hardness, and impact strength are directly affected. Secondly, as the hardening penetration depends on the size of the carbides, the complex properties, such as hardenability and machinability, are also affected. (2) In order to determine the size of the carbides a formula similar to that used for the definition of grain size is applied. The formula is $M = 2^N$ (where M is the number of particles in thousands/sq.cm. and N is the dispersion. The dispersion N and hardness remain in a straight-line relationship which is shown in a dispersion/Brinell-hardness diagram. (3) The mechanism of the tempering of martensite, from the point of view of the morphology of cementite, is revealed by a growth of the cementite spheroids with time and temperature. (4) The lamellar structures are essentially unstable and tend to become spheroids. In the course of this evolution each lamella contracts into grains the diameter of which, at the end of the period of the complete disintegration, is 15 to 16 times the width of the lamella. (5) The evolution of pearlite into a globular form is very slow and is effected by annealing. (6) The coarse carbides act, similarly to other inclusions, as nuclei for crystallization; they affect the structure and quality of steel by influencing its hardenability. They impede the grain growth, at the same time preventing overheating, and they reduce the hardenability. On the other hand, the fine carbides easily enter into solution and their presence is indispensable to obtain the hardening effect in steel after quenching. (7) The coarse carbides in the hypereutectoid carbon steels increase their cutting ability and produce the same effect as the complex cementites in the high-speed tool steels. (8) Recent investigations on grain size and hardenability, which are properties closely connected with the distribution, form, and size of carbides, made it possible in many cases to replace high-alloy steels with plain carbon steels.

Some Factors Affecting Edgewise Growth of Pearlite. W. H. Brandt. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1857: Metals Technology, 1945, vol. 12, Dec.). In an earlier paper (*see* Journ. I. and S.I., 1945, No. II., p. 97 A) the author presented a method for calculating V, the velocity of the edgewise growth of pearlite, using linear extrapolations of the ferrite and cementite solubilities. Whilst the results were in fairly good agreement with the experimental data, the calculated velocity of growth at 600° C. was, contrary to expectation, smaller than G (the velocity of radial growth of pearlite nodules). In this paper the calculation is repeated using a new extrapolation for the ferrite solubility. A calculation is also made of the effect of the carbon content on the value of V. The problem of calculating the effects of other alloying elements on rates of growth is also dealt with.

Equilibrium Relations in Medium-Alloy Steels. C. Zener. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1856: Metals Technology, 1946, vol. 13, Jan.). The general theory for the equilibrium between two solid phases is developed from thermodynamic and mechanical concepts. This general theory is applied to the equilibrium relationships between the α, γ, and cementite phases of the iron–carbon system, and to the effects of alloying elements thereon. Experimentally obtained data on equilibria are brought into a consistent picture and a method is developed for rapidly computing the equilibrium relationships in any medium-alloy steel.

Phase Boundaries in Medium-Alloy Steels. W. A. West. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1924:

Metals Technology, 1946, vol. 13, Jan.). Equations have been developed by Zener (*see* preceding abstract) which permit the calculation of phase boundaries in medium-alloy steels in terms of certain constants. In this paper experimental data from the literature have been used to determine the constants for a number of alloying elements and these have been compared with the constants obtained by Zener's equations. Two systems have been found which do not conform with Zener's theory; in certain others the data were inconclusive or insufficient; in a considerable number, however, constants may be chosen that represent the experimental results reasonably closely.

Kinetics of the Decomposition of Austenite. C. Zener. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1925: Metals Technology, 1946, vol. 13, Jan.). A review is made of the general principles that govern the phenomena accompanying the decomposition of austenite; these include nucleation, propagation of interfaces, grain growth, and spheroidization. The main features of pearlite formation can be derived directly from fundamental principles. The formation of bainite is interpreted as a transformation of the face-centred austenite lattice to the body-centred ferrite lattice without a change in carbon distribution. Martensite is interpreted as having essentially the same structure as untempered bainite, except that, owing to its mode of formation, residual stresses are present. The manner in which alloying elements affect the formation of martensite is found to be such that, when arranged in the order in which they lower the Ar″ point, the order is the same as when arranged according to effectiveness in retarding the bainite transformation. The transformation of martensite from a tetragonal lattice to a cubic lattice is interpreted as a change from an ordered distribution of carbon to a random distribution.

BOOK NOTICES

(Continued from p. 71 A)

BENNETT, H. "*Substitutes.*" 8vo, pp. x + 225. Brooklyn, N.Y., 1943: Chemical Publishing Co., Inc. (Price 21*s.*)

A handbook of substitutes and alternatives for chemicals, metals, fibres, and other commercial products, including a plan for choosing proper substitutes.

BOSTON, O. W. "*A Bibliography on Cutting of Metals 1864–1943.*" 8vo, pp. xi + 547. New York, 1945: American Society of Mechanical Engineers. (Price 39*s.*)

Professor Boston has already prepared two volumes of references—one published as a Research Bulletin of the A.S.M.E. in 1930, and a second one published personally by him in 1935. The present volume consists of the material published in the two previous volumes, together with nearly 3500 additional references with abstracts; it forms the most complete bibliography of the literature on the cutting of metals.

DRINKER, P., and T. HATCH. "*Industrial Dust. Hygienic Significance, Measurement, and Control.*" 8vo, pp. viii + 316. Illustrated. New York, 1936: McGraw-Hill Book Co., Inc.; London: McGraw-Hill Publishing Co., Ltd. (Price 24*s.*)

This book, which is written from the engineer's standpoint, contains the following chapters: (1) Physical properties of dusts, fumes, and mists; (2) Effects of dusts and fumes upon man; (3) Physical and chemical factors in pneumoconiosis; (4) Permissible dustiness; (5) The dust survey; (6) and (7) Dust determination; (8) Determination of particle size; (9) Chemical and minerological analyses; (10) Control of the dust hazard—General; (11) Design of local exhaust systems; (12) Air-cleaning apparatus; (13) Air-cleaning apparatus—Filtration; (14) Dust respirators and air masks.

EVANS, E. A. "*Lubricating and Allied Oils.*" Foreword by the late Viscount Wakefield of Hythe. Third edition, revised. 8vo, pp. ix + 210. Illustrated. London, 1945: Chapman and Hall, Ltd. (Price 15*s.*)

This new edition has been almost completely rewritten, whilst its original scope and outlook have been preserved. Several new chapters have been added which, it is hoped, will increase the usefulness of the book. Fatty oils have been described in a way which should give chemists a better appreciation of their constitution and behaviour. The question of engine deposits is also discussed and an introductory survey made of the new subject of chemical addition agents.

HALL, J. J. "*Steel Hardening, Tempering, and Annealing.*" Sm. 8vo, pp. 160. Illustrated. London, 1945: George Newnes, Ltd. (Price 6*s.*)

This book, which gives a concise account of the theory and practice of the heat-treatment of steel, is intended for tool hardeners, mechanics, and others interested in the manufacture and maintenance of cutting tools, dies, and other engineering components requiring heat-treatment.

JARVIS, E. G., and H. O. JARVIS. "*Facts for Foundrymen.*" Sm. 8vo, pp. ix + 80. Buffalo, N.Y., 1944: Niagara Falls Smelting and Refining Corporation. (Price $1.00.)

This small book outlines the properties and uses of the most commonly used metals and alloys. The information is presented for the foundryman to assist him in the production of better castings.

MAY, E. C. "*Principio to Wheeling 1715–1945. A Pageant of Iron and Steel.*" 8vo, pp. xiv + 335. New York and London, 1945: Harper and Brothers. (Price 15*s.*)

This volume traces the evolution of a colonial iron furnace at Principio, Maryland, into the huge blast-furnace, open-hearth, and rolling-mill plant at Wheeling, West Virginia, and describes the present-day activities of the Wheeling Steel Corporation.

NATIONAL RESEARCH COUNCIL. "*Chemistry of Coal Utilization.*" H. H. Lowry, Editor. 8vo. Vol. 1, pp. cv + 920. Vol. 2, pp. cv + 921–1868. Illustrated. New York, 1945: John Wiley and Sons, Inc.; London: Chapman and Hall, Ltd. (Price £6 the 2 volumes.)

This treatise constitutes a comprehensive and critical review of the voluminous but scattered literature on the

scientific and practical aspects of coal utilization. In many cases hitherto unpublished data have been included. The material is presented in two volumes.

Vol. I. comprises 24 chapters covering, among other subjects, the origin of coal; its classification; its petrography; its calorific value; its hardness, strength, and grindability; its general physical properties; its chemical constitution; the fusion, flow, and clinkering of coal ash; the cleaning of coal; changes that take place in coal on storage; the action of solvents on coal; vacuum distillation of coal; coal carbonization; and coke.

Vol. II. comprises 16 chapters dealing with such subjects as the recovery of sulphur and nitrogenous compounds from coal gas; light oil from coke-oven gas; removal of miscellaneous constituents from coal gas; utilization of coal gas; the chemical nature of coal tar; ammoniacal liquor; combustion; direct generation of electricity from coal and gas; water gas; hydrogenation of coal and tar; synthesis of hydrocarbons and of methanol from water gas.

REICH, H. J. "*Principles of Electron Tubes.*" 8vo, pp. xv + 398. Illustrated. New York, 1941: McGraw-Hill Book Co., Inc.; London: McGraw-Hill Publishing Co., Ltd. (Price 21*s.*)

This book has been specially prepared to meet the needs of students who do not intend to specialize in communication. It is an abridgement of a previous work entitled "Theory and Applications of Electron Tubes."

STEWART, F. S. "*Airframe Materials.*" 8vo, pp. x + 237. Illustrated. New York, 1945: McGraw-Hill Book Co., Inc.; London: McGraw-Hill Publishing Co., Ltd. (Price 15*s.*)

It contains the following chapters: (1) The sources of strength of materials; (2) Properties of materials and methods of testing for quality; (3) Comparison of certain physical and mechanical properties of airframe materials; (4) Heat-treatment of aluminium alloys; (5) Heat-treatment of steel, magnesium, and other materials; (6) Effect of cold-working on the properties of aluminium alloys; bending limitation of wood, plastics, and magnesium; (7) Corrosion of metals and ageing of plastics; (8) Cleaning and chemical treatment; (9) Bonding processes used to join materials structurally; (10) Casting and forging of aluminium and magnesium; other metals; (11) Veneers and plywood. Adhesives; (12) Sealing processes used in airplanes; problems involved; (13) Plastic compositions; (14) Transparent plastics.

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RITCHEY, J. "*Pattern Making.*" Revised by W. W. Monroe, C. W. Beese, and P. R. Hall. Pp. 233. Chicago: American Technical Society. (Price $2.00.)

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